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MATERIALS SCIENCE

Processes in Materials Science

Additive Manufacturing of Metal Alloys 1

*Processes, Raw Materials
and Numerical Simulation*

**Coordinated by
Patrice Peyre
Éric Charkaluk**

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Materials Science, Field Director – Jean-Pierre Chevalier

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Introduction

Patrice PEYRE¹ and Éric CHARKALUK²

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Fuelled by Barack Obama's 2013 State of the Union speech (Elyan and IDG NS 2013), which highlighted 3D printing as the spearhead of the manufacturing industry of the future, additive manufacturing (AM) has experienced an unprecedented craze and exponential growth for several years, as have associated scientific and technological developments. According to the *New York Times*, this is indeed the industrial revolution 2.0, offering the possibility of printing personalized parts and small and medium series at increasingly lower costs.

It has been a long time since the first patent was filed for additive manufacturing in 1984 by the French trio J.-C. André, O. De Witte and A. Le Méhauté (André 2017). In 10 years (2007–2017), the annual revenues generated by additive manufacturing have increased eightfold and the number of scientific publications on the subject by five. Not a week goes by without the press reporting a revolutionary new additive technology, a new material being developed or a new field of application being explored. All of this goes well beyond the fad: witness the considerable success of the last Formnext face-to-face show on the subject (Frankfurt, November 23–27, 2019), reflecting the growing maturity of current processes and the emergence of different derived processes.

The short-term prospects for different ranges and sizes of parts are also exciting, from those associated with bioprinting, allowing the printing of blood vessels or human skin, to the construction of individual houses by printing layers of concrete.

This new manufacturing paradigm directly impacts a large number of sectors such as raw materials (powders, wires), manufacturing machines (3D printers, material deposition machines, etc.), 3D modeling software (topological optimization) and production systems (lasers, non-destructive testing, real-time control systems, etc.), to name a few (Heinrich 2013). For more information on the technical and economic aspects associated with additive manufacturing, the reader may refer, for example, to PIPAME publications on the subject (PIPAME 2017).

One of the strengths of additive manufacturing lies in its incredible power to attract the younger generations, who find it a good way to express themselves when participating in the design or manufacture of parts of formidable complexity, often inspired by nature and so far unachievable with traditional technologies. The printing of three-dimensional parts is therefore part of our daily life and calls on the imagination as much as it responds to everyday problems (manufacturing of spare parts for example).

It is therefore easy to understand that the engineer, the technician and the project manager of the future factory must control the entire value chain of additive manufacturing at different levels, since the development of this new creativity unleashes the imagination of designers and even allows one to predict the sustainability of manufactured parts through the appropriation of associated processes.

In this very exciting and competitive context (in France, each region has launched its own initiatives on the subject), writing a book on metal additive manufacturing is undoubtedly necessary, but also a challenge as technology is evolving so rapidly, driven by unprecedented industrial challenges. The risk of planned obsolescence is therefore real, even if, in the field of metallic materials, certain technologies have already reached a real degree of maturity.

More concretely, the production of small and medium series of “acceptable” metallic parts (density close to 1) by additive manufacturing is now possible using different types of processes, which are mainly distinguished by differences in complexity or dimensions of achievable parts. On this point, 5 years ago, the largest parts achievable using powder laser bed fusion (L-PBF) did not exceed 40 cm, whereas they are now more than double this.

In addition, new processes appear regularly. To name just one, we can mention the recent patent (2017) filed by P. Teulet (formerly of the company Phénix) for additive micro welding in which material is manufactured by the laser welding of thin metal bands layer by layer (Teulet 2017).

The objective in the following pages is therefore to establish the least obsolete state of the art possible on the subject in 2021 by focusing on metallic materials, their different development conditions and the basic physical principles involved, then concluding with the microstructures and the material properties of the produced parts. There is no historical review as there is little or no notion of creativity in additive manufacturing, bio-inspired design and topological optimization, but there is a well-assumed process-materials approach, and an important part dedicated to numerical simulation, the latter being essential in the detailed physical understanding of the complex processes involved. For more information on additive manufacturing, design rules and standardization, readers are invited to consult the recent general work by Barlier and Bernard (2020).

The chapters that make up this work were originally written in French by recognized French-speaking specialists, with a balance, for each topic, between fundamental and more advanced aspects, directly derived from laboratories and associated manufacturers.

This work consists of two volumes:

- this book, Volume 1, presents the metallic AM processes, the raw materials, the physics of AM processes and numerical simulation;
- Volume 2 (Peyre and Charkaluk forthcoming) presents the microstructures, post-processing and material properties of AM parts.

All this is intended to form an overview of metal additive manufacturing which, we hope, will be useful to as many people as possible.

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1

Metal Additive Manufacturing Processes

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According to the French standard NF E 67-001, additive manufacturing is the set of processes allowing the layer-by-layer manufacture of a physical object by adding material from a digital object. Within these processes, metallic materials occupy an increasingly important place due to their growing maturity, but also due to the industrial desire to move from mock-up to direct manufacture of parts.

In this first chapter, we present the main additive processes currently used at the industrial level for the manufacture or repair of metal parts. We discuss the basic principles, the main process parameters and their effects on the geometries of the parts, as well as various technological and industrial considerations. Additional information is presented in Chapter 3 on the physics of metal additive manufacturing processes.

Before discussing these different metal additive manufacturing processes, it is advisable to briefly recall the different stages of the manufacture of a part, common to any additive process (Figure 1.1). The part to be produced is digitized using computer-aided design (CAD) software, the shape of which most often results from further processing by topological optimization (TO) software. The digitized model of the part is then converted into the computer language of the machine (.stl file). This file is then processed with CAD/computer-aided manufacturing (CAM) software, which allows the part to be split into successive layers. Each layer is assigned a set of “machine” parameters and “manufacturing” parameters, which must minimize

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thermal gradients as much as possible and thus reduce the internal stresses of the part. Mastery of the thermal–metallurgical–mechanical trio is therefore at the heart of scientific issues (Figure 1.2), as for all thermal processes. Furthermore, it is necessary to install supports below certain undercut surfaces and on very thin areas, so as to avoid deformation of the part during its construction. This is really the step that concentrates all the know-how of the manufacturers of 3D printed parts. At the end of the manufacturing process, the supports are removed before the part undergoes geometric, dimensional and metallurgical quality control checks. After successfully passing these various checks, post-processing may possibly be considered and may consist of both heat treatments and surface finishing treatments (see Chapter 2 of Volume 2).



Figure 1.1. Main stages of additive manufacturing

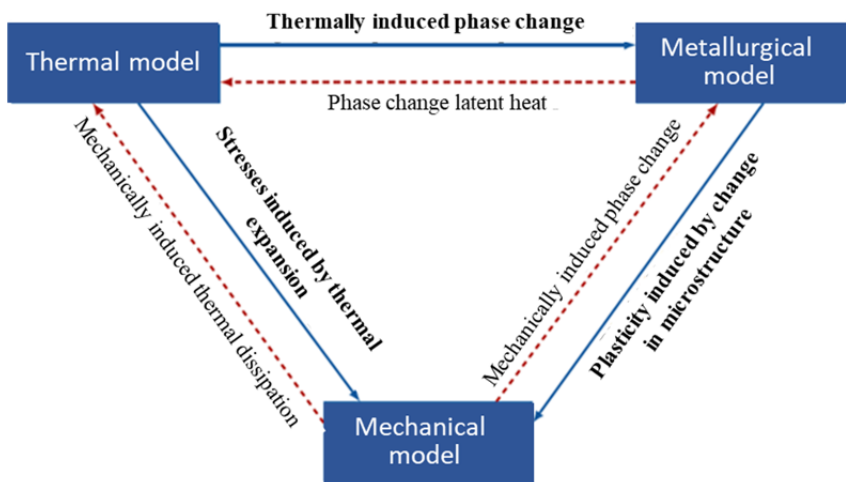


Figure 1.2. Representation of interactions between thermal, metallurgical and mechanical models, allowing a better understanding and the numerical simulation of the physical phenomena leading to the establishment of internal stresses in raw additive manufacturing parts (Marion 2016)

We can usually classify the additive manufacturing processes according to seven main families, as described in Table 1.1.

The following sections therefore review the essentials of the metal additive processes most widely used to build dense 3D metal materials. The chapter concludes with some upcoming processes, the list of which is far from exhaustive due to the continuous and regular emergence of new concepts.

	Thermosetting polymers	Thermoplastic polymers	Wood	Metals	Industrial ceramics	Functional ceramics
Vat photopolymerization						
Material projection						
Binder projection						
Powder bed sintering/fusion						
Material extrusion						
Directed energy deposition						
Stratification						

Table 1.1. Additive manufacturing processes (AFNOR 2017). For a color version of this table, see www.iste.co.uk/peyre/alloys1.zip

1.1. The DED-LMD process

1.1.1. Process overview

The laser metal deposition (LMD) process can be considered the oldest additive manufacturing process, as it is directly derived from laser cladding processes (surface treatments using material deposition) developed in the 1980s, at the same time as the dawn of industrial laser processes (welding, cutting, etc.). One of the most well-known applications of LMD is the deposition of Stellite® (a wear-resistant cobalt-based alloy) on steel valve seats. The shift from 2D applications (surface treatments) to 3D applications occurred in the early 2000s with the development of industrial machines. As it stands, the recognized advantages of the process are as follows: (1) the low rate of residual porosity, (2) the

good metallurgical continuity between layers, (3) the ease of automation and control of the process, (4) the possibility of use on a wide range of materials and (5) the possibility of developing materials with chemical composition gradients. It is mainly used today to repair (or reload) damaged or rejected parts at high cost.

1.1.2. Basic elements

LMD is part of the directed energy deposition (DED) process according to the standard ASTM 52900 (2015), as well as wire arc additive manufacturing (WAAM) and wire and laser additive manufacturing (WLAM). As such, it involves depositing successive layers via a coaxial head (Figure 1.3), ensuring both the projection of powder via a dedicated projection nozzle and the focusing of the laser due to beam collimation and focusing, which allows the image of the optical fiber to be outputted on the work plane, multiplied by a magnification factor depending on the focal length ratio $= F_{\text{foc}}/F_{\text{coll}}$. In its conventional configuration, the focal plane of the powder jet coincides with the focal plane of the laser, which generally allows for optimized interaction efficiencies.

Physically, the projected powder heats up during its time of flight when the powder particles enter the laser beam caustic (the envelope of the focusing light rays), and it can change to the liquid state provided that the powder/laser interaction distance is long enough to accumulate the energy ($E = m_p \cdot C_p \cdot (T_F - T_0) + m \cdot L_F$) necessary for particle fusion (with a typical diameter between 45 and 125 μm). In the conventional configuration, the nozzle–substrate distances are relatively short (between 3 and 10 mm in most cases) and the laser heats the powder particles without melting them (Figure 1.3). The latter will therefore melt on contact with the fusion zone (FZ) generated by the laser in the substrate before being stirred by the fluid movements within the FZ (hydrodynamics of the melt-pool), induced by the thermocapillary convection (Marangoni flow) (see Chapters 3 and 4 of this volume). Experiments with laser transmission through the powder jet (Peyre et al. 2008) estimate the proportion of energy absorbed in the powder to be between 5% and 20% and between 80% and 95% in the substrate (the previous layer) in usual deposit conditions. Ultimately, the process interaction efficiency corresponds to the proportion of the powder jet included in the FZ and therefore the ratio between the mass of powder melted and incorporated in the part and the mass projected. This η_p ratio, also called the mass melting efficiency, is generally around 50–80% and, with an equivalent powder jet, increases with the size of the FZ.

For energy efficiency reasons, the lasers used are most often solid continuous lasers of the type Nd:YAG or Yb:YAG with wavelengths of 1.03–1.07 μm or laser diodes emitting power around 0.8 μm . Pulsed regimes, although less widely used, can help improve the surface finish of manufactured parts but lower the mass yield.

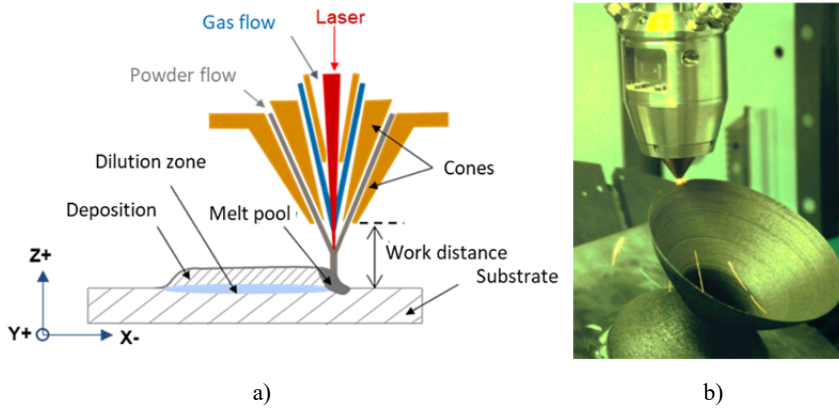


Figure 1.3. a) Principle of the LMD process; b) part being manufactured (Beam 2020). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

Powder distributors bring the powder to the nozzle and can be of different types. One of the most common is a rotating circular plate containing a groove in which the powder flows by gravity from a conical bowl in which it is stirred to avoid agglomerates (Figure 1.4(a)). The plate undergoes a complete rotation cycle before the powder is brought by the carrier gas (argon), at speeds of a few m/s. The rotational velocity of the plate then conditions the mass flow (between 1 and $25 \text{ g} \cdot \text{min}^{-1}$) at the nozzle outlet.

The projection nozzles (using copper-based alloys with high thermal conductivity to avoid heating by radiation from the FZ, see Figure 1.4(b)) generally consist of two to four concentric cones between which pass (1) the carrier gas that transports powder, (2) a central gas that protects the optics and (3) an external gas that improves protection against oxidation. This results in a powder jet with a quasi-Gaussian distribution D_m^* at the powder focal plane (Figure 1.5, equation [1.1]) and annular outside this plane. As such, the spray nozzle is the most essential element of the process, the design of which allows the diameter at the focal point of the jet to be changed (typically between 0.8 and 4 mm) as well as the nozzle-substrate distance. Industrially, we often try to increase this distance to avoid overheating that can contribute to degradation or clogging of the nozzles:

$$D_m^* [\text{g/s/m}^2] = n \frac{D_m}{\pi r_{\text{jet}}^2} \exp\left(-n \frac{(x^2 + y^2)}{r_{\text{jet}}^2}\right) \quad [1.1]$$

where D_m = average mass flow ($\text{g} \cdot \text{min}^{-1}$), r_{jet} = powder jet radius (m) and n = jet shape adjustment parameter (generally between 2 and 5).

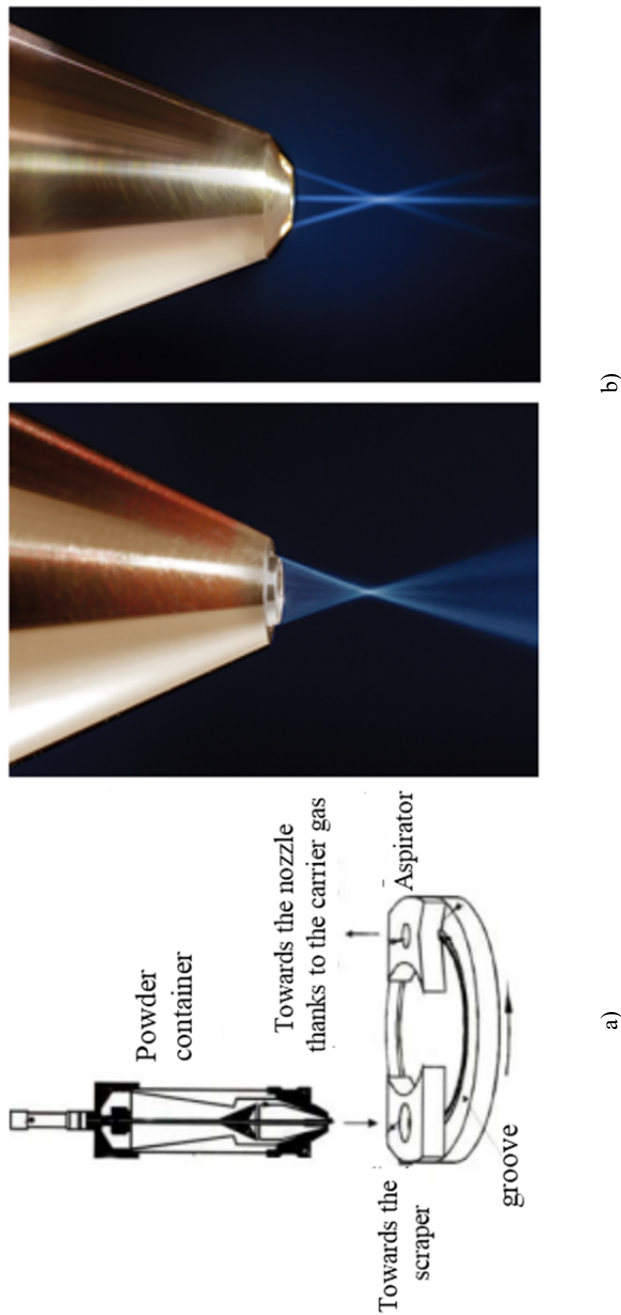


Figure 1.4. a) Principle of a rotary distributor; b) examples of single-jet and three-jet nozzles (Mezari 2014)

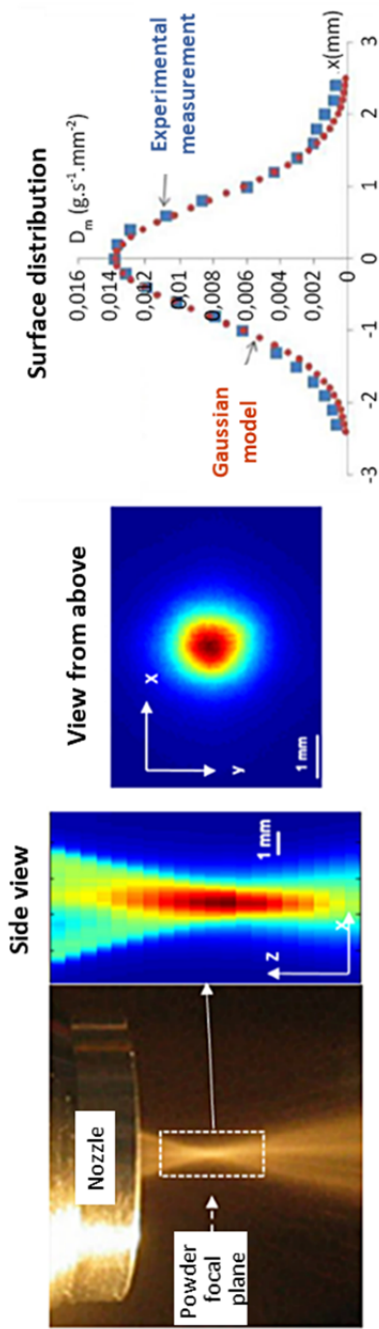


Figure 1.5. Image of the LMD powder jet, associated caustic (measurement by laser sheet plus camera) and area distribution of mass flow, based on Gharbi (2013). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

In the end, LMD machines fall into two categories: those with a 5-axis installation with or without overall gas protection, and those for which a robot arm moves the projection head relative to the part. In both cases, the projection axis must be as perpendicular as possible to the part during the process in order to avoid the collapse by gravity of the liquid metal.

1.1.3. Overview of process parameters and their influence

1.1.3.1. The different process parameters

The parameters of the LMD process can be distinguished into two categories: “machine” parameters often kept constant for a given application and so-called “manufacturing” parameters, on which the quantity of material and energy deposited on the part depend and which enter directly into the initial manufacturing program. These parameters are summarized in Figure 1.6.

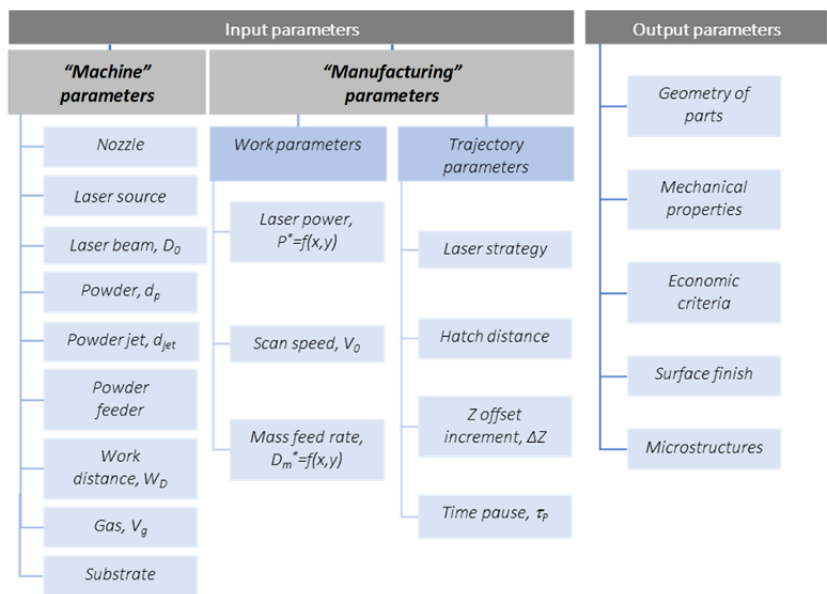


Figure 1.6. Input and output parameters of the LMD process (Ferreira 2021)

The *manufacturing parameters* (laser power P_0 (between 200 and 3,000 W), relative nozzle/substrate advance velocity V_0 (between 0.001 and 0.030 m·s⁻¹), mass flow rate D_m (from 1 to 25 g·min⁻¹), spatial distribution of laser power $P^* = f(x,y)$ and spatial distribution of powder jet mass flow $D_m^* = f(x,y)$) condition the

dimensions of the LMD beads and more generally those of the manufactured parts. The scanning speed is, however, a parameter that should be considered with caution, because it conditions both the supply of material and the laser energy.

The *machine parameters* include the laser diameter D_0 (from 0.6 to 4 mm), laser/powder defocusing, the nature (Ar, N₂) and velocity of the various gases integrated in the nozzle, the size distribution of the powder particles, etc. All of them also play a role in the stability and yield of the process.

Finally, the *trajectory parameters* condition the optimal overlap between single tracks and the controlled distribution of heat in the room in order to obtain the best final properties (high density, reduced residual stresses). They include the hatch, that is, the inter-track distance, the inter-track pause time τ_p (s), the scanning method (one-directional, two directional, in-line, spiral, etc.) and the nozzle Z offset increments (ΔZ).

1.1.3.2. Influence of process parameters on the geometry of manufactured parts

The combination of laser and powder parameters directly depends on the dimensions of the manufactured parts. Unit bead widths manufactured in LMD are between 0.5 and 5 mm, depending on the parameters (laser power, scanning velocity, beam diameter) used. These widths delimit the parts manufactured in the case of wall-type structures (for example the reloading of a guide vane). Contrary to powder bed processes for which the layer height Δh is programmed via the incremental descent of the production plates, the LMD layer heights (between 0.1 and 2 mm) depend entirely on the process parameters (P_0 , V_0 , D_m) and may change during manufacture depending on the powder/FZ coupling. Thus, the first layer formed directly on the substrate is often narrower and higher because the FZ is centered on the top of the Gaussian distribution of the powder jet.

The evolution of these two values (width and height) as a function of the primary parameters of the LMD process is presented in Figure 1.7. There are several important aspects to note: (1) wall widths increase with power and decrease with velocity, which reflects the linear energy effect $E_l = P_0/V_0$ (J·m⁻¹) in the widening of the FZs, (2) the layer heights increase with D_m and with the inverse of the velocity $1/V_0$, which reflects the effect of powder accumulation per bead length D_m/V_0 (kg·m⁻¹). The latter saturate beyond a certain mass flow threshold corresponding to the balance between the energy required to melt the projected particles and that contained in the metallic melt-pool. Finally, the wetting angle θ (Figures 1.8 and 1.9) decreases with energy intake. During parametric optimization, we try to generate angles θ less than 80° to promote lateral overlap and avoid lacks of fusion.

Different empirical analytical formulations make it possible to relate geometric quantities to process parameters (Balit 2019). These formulations take the following form: $\Delta h = P_0^\alpha \cdot V_0^\beta \cdot D_m^\gamma$ with three indices α, β and γ , which vary greatly depending on the authors, the materials and the experimental conditions ($\alpha, \gamma > 0$ and $\beta < 0$). Charts can also be produced to help users choose the primary parameters of the process (P_0, V_0, D_m) considering criteria such as bead height Δh , bead width e , the height of dilution h_{dil} , the apparent transverse cross-section of the bead S_{app} , mass yield η_p and the wetting angle θ (Figure 1.9). The parts manufactured by the LMD process are of moderate complexity compared to a powder bed process such as laser powder bed fusion (L-PBF) (Figure 1.10). Considering a simplified calculation of the hourly production volume flow ($D_v = \Delta h \cdot e_v \cdot V$, with $e_v = \text{hatch}$), parts are produced much faster ($100\text{--}300 \text{ cm}^3 \cdot \text{h}^{-1}$ vs. $20\text{--}50 \text{ cm}^3 \cdot \text{h}^{-1}$) than those made with the L-PBF process, mainly due to the unitary fusion beads being 10–20 times higher and 10 times wider for a scan velocity 20–100 times slower.

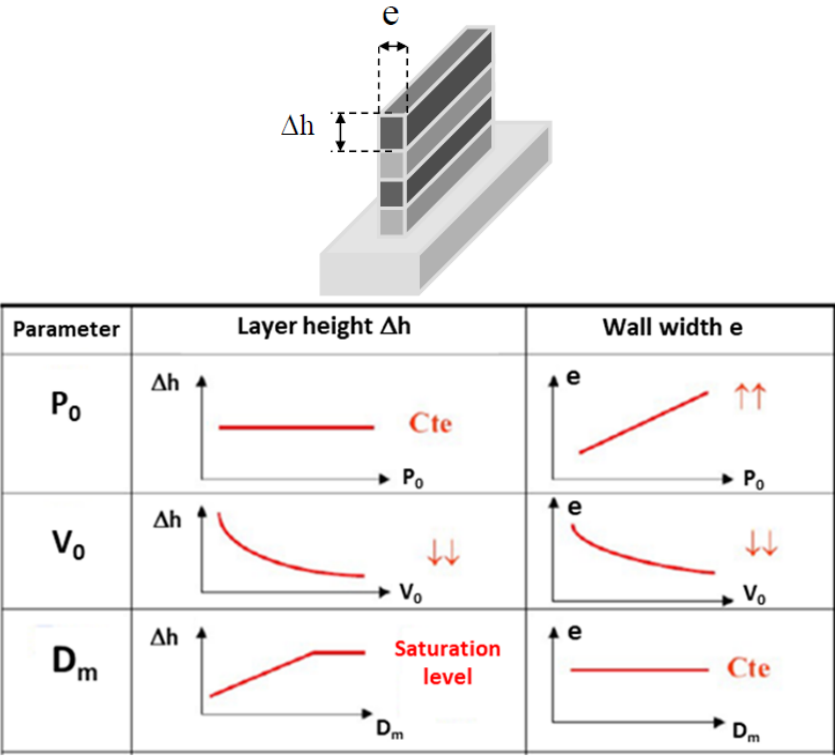


Figure 1.7. Effects of parameters P_0, V_0 and D_m on LMD wall geometry (Maisonneuve 2008)

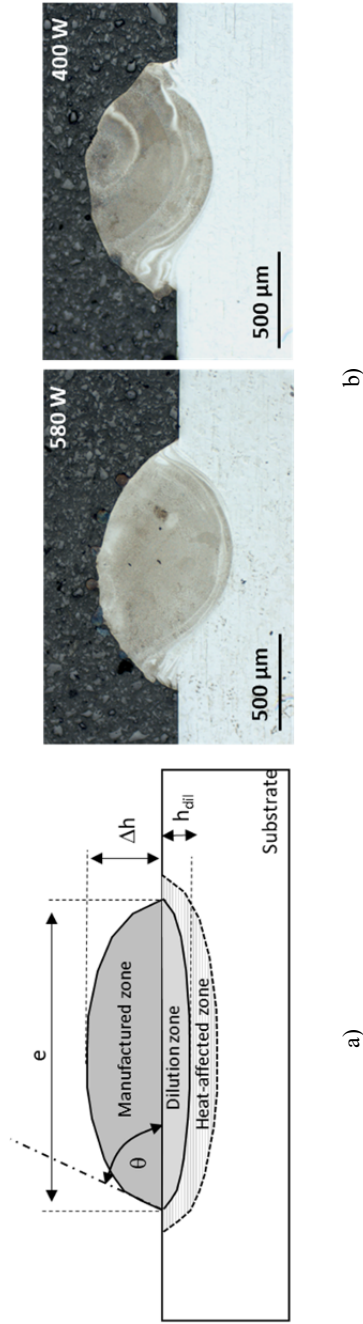


Figure 1.8. LMD mono-beads: a) characteristic zones; b) martensitic stainless steel beads made in 316L (when $P0 \searrow$, Δh varies little but $V_{melted} \searrow$)

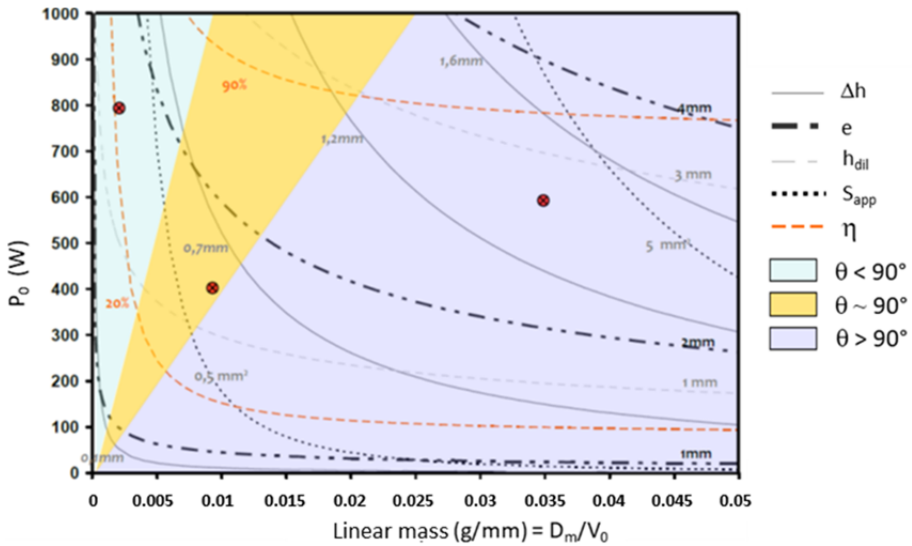


Figure 1.9. Process mapping: main geometric variables of an LMD bead as a function of the laser power P_0 and linear mass D_m/V_0 (Maisonneuve 2008). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

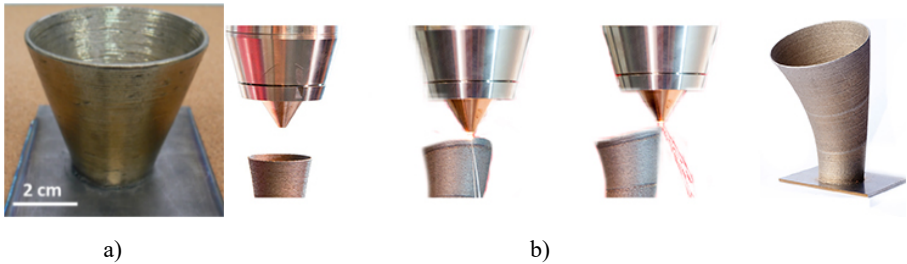


Figure 1.10. Construction of a 3D part by LMD: a) conical titanium matrix composite part (Pouzet 2015); b) reloading of a part (Beam 2020)

1.1.3.3. Surface conditions of LMD parts

In most cases, due to the relatively poor quality of the surface conditions obtained on LMD blanks ($R_a > 15 \mu\text{m}$ on average), the latter are re-machined to meet the standards of the targeted applications. These surface conditions come from two origins: (1) the discretization in layers of 0.1–1 mm, which leads to the formation of lateral menisci (the liquid metal collapses under the effect of its own weight despite the presence of surface tensions); (2) the agglomeration of projected

powder particles unmelted on the still hot surfaces (Figure 1.11). Improvements in surface finishes are possible with the use of a pulsed laser regime, but at the expense of process productivity (Gharbi 2013).

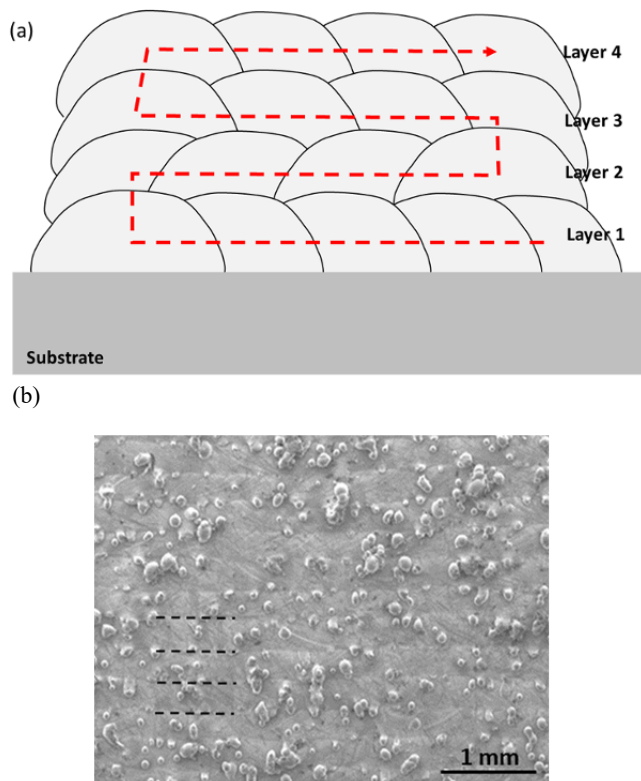


Figure 1.11. a) LMD construction of 3D parts; b) SEM image of an LMD surface (agglomerated particles and superimposed layers in dotted lines)

1.1.4. Thermal cycles induced by the process

As with most thermal additive processes, the LMD process induces a thermal cycle made up of at least two fusion-solidification cycles (four fusion cycles in Figure 1.12), with cooling rates of the order of $100\text{--}2,000\text{ K}\cdot\text{s}^{-1}$ and laser-matter interaction times ($\sim D_0/V_0$) from 0.1 to 4 s.

These thermal conditions, as well as the size of the beads formed, are similar to those of a conduction laser welding process (almost without vaporization) followed by a process of tempering the consolidated beads over the entire manufacturing period.

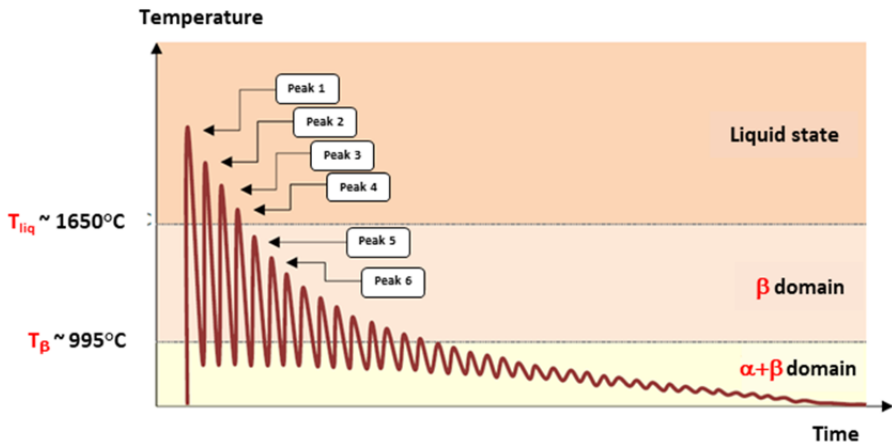


Figure 1.12. Point temperature within a Ti-6Al-4V LMD wall produced by LMD after the deposition of 40 additional layers (Blanc 2017)

Let us return in more detail to the thermal history of a layer within a single-bead wall made of Ti-6Al-4V, as shown in Figure 1.12. The construction of a layer n of such a wall requires a first temperature peak above the liquidus temperature of the material ($T_f \sim 1,650^{\circ}\text{C}$). This first melt-pool solidifies rapidly under near out-of-equilibrium conditions to an annealing temperature after the laser beam has moved away from the temperature measurement point. The deposition of a second layer $n + 1$ brings the material of the previous layer to a lower heating rate and to a lower peak temperature than that of the first peak, whether or not accompanied by melting, depending on the operating conditions used. Cooling then takes place at a slightly slower rate and conduces to the same annealing temperature. Each additional layer deposition induces a new heating–cooling cycle. From one layer to another, as the temperature peak decreases much faster than the annealing temperature, the amplitude of the thermal oscillations attenuates to become negligible after a large number of deposited layers. Thermal cycling then ceases and the material slowly cools down to room temperature. This thermal story of a single-bead wall shows that for a layer-by-layer construction of a part of any shape and whatever the scanning method used, the temperature at each point of the part changes anisothermally. This thermal cycling in fields β and $\alpha + \beta$ of Ti-6Al-4V will induce (i) significant metallurgical changes of different natures depending on transformation temperatures and cooling rates and (ii) more or less significant relaxations of internal stresses depending on the equivalent time spent in the solid state (see Chapter 4 of this volume).

1.1.5. Types of materials involved

Most metallic materials can be sprayed in powder form and used in LMD, with the exception of a few limitations for metals with high reflectivity R under laser illumination and high thermal conductivity. Thus, aluminum alloys are rarely used ($R > 80\%$), nor copper alloys ($R > 90\%$). In addition, the materials considered to crack during welding (the case with nickel alloys with a high rate of precipitates γ') cause the same type which problem in LMD. The cohesive nature of certain powders (Touze 2019) promoting agglomerates and limiting flow through distribution systems can also be problematic (see Chapter 2 of this volume).

1.1.6. Microstructures of manufactured or repaired parts

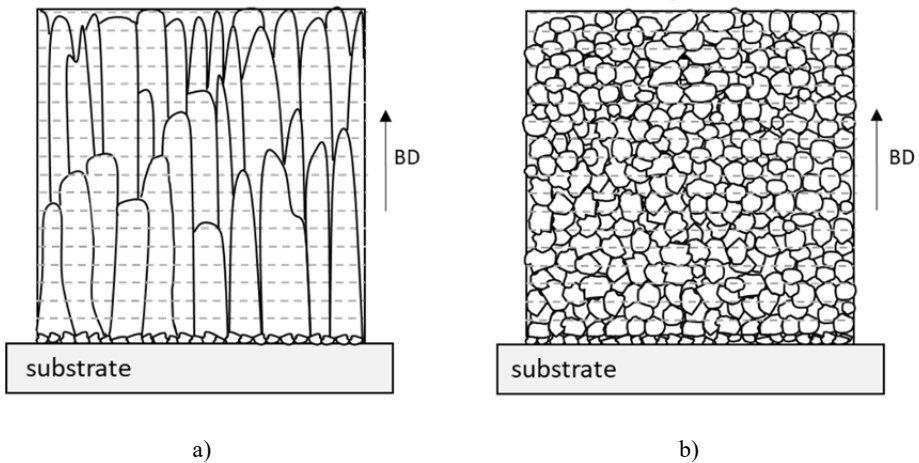


Figure 1.13. (a) Columnar and (b) equiaxed microstructures obtained in LMD (BD = build direction)

The grains resulting from LMD solidification can be columnar (oriented according to the build direction) or equiaxed (no preferential orientation) depending on the process parameters and, to a lesser degree, the position in the parts (Figure 1.13). A columnar morphology will be favored by a high linear energy P_0/V_0 ($J \cdot m^{-1}$) resulting in a high interlayer dilution rate, high thermal gradients and low solidification rates. An equiaxed morphology will be possible when the remelting of the previous layers is minimized (case of low linear energies P_0/V_0), the liquid metal solidifying almost autonomously like a droplet placed on the previous layer. Unlike

columnar morphology, thermal gradients are low and solidification rates high for equiaxed morphology. In addition, certain singular zones (wall/deposit interface) favor a less oriented heat dissipation and can lead to equiaxed grains. At a finer scale (inside metallurgical grains), the microstructure can be dendritic or cellular depending on the local thermal cycle (see Chapter 1 of Volume 2).

Certain solidification structures can also present atypical morphologies (Figure 1.14), which reflect a competition between different heat flow modes (vertically/horizontally) and the influence of the geometry of the LMD fusion beads.

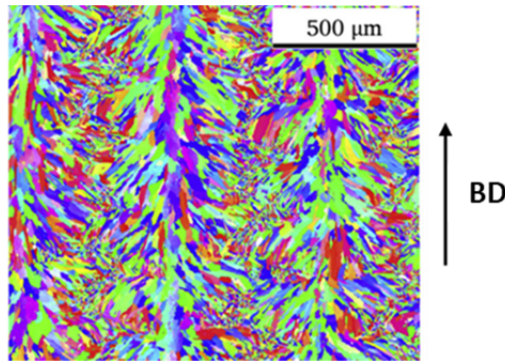


Figure 1.14. Microstructure obtained by LMD on a 3D part in Inconel 625 with V-shaped solidification structure (Nguejio et al. 2019). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

1.1.7. Industrial systems

Two types of industrial systems are available on the market, and they depend essentially on how the powder nozzle is moved relative to the substrate and on the presence of an inerted enclosure (Figure 1.15).

This last aspect is particularly important for materials reactive to O_2 such as titanium alloys, the ductility of which drops rapidly with oxygen contamination (above 2000 ppm according to aeronautical standards). For other types of materials, local protection against oxidation, via inert gases from the powder injection nozzle, is often sufficient.



Figure 1.15. Example of an LMD industrial installation (BeAM machine)

1.1.7.1. Kinematics and technological constraints

The nozzle-part positioning is an important parameter in the process. To build or repair the part, the machine deposits the material following the programmed trajectories, layer after layer. The required positioning accuracy depends on the characteristics of the process implemented. Thus, with LMD processes, the beads of deposited material can be thin (1–3 mm), with a stable and reproducible geometry. In this case, Cartesian systems are often preferred because they allow more precision. By way of comparison, the wire deposition processes (WAAM), allowing higher manufacturing speeds than LMD, wider beads (8–20 mm) and less stable geometry, use anthropomorphic systems because they are less costly, with lower precision.

The growth direction (stacking direction of layers) of the part is often vertical. First, this allows the layers to be stacked while limiting the effects of gravity on the spreading of the liquid metal layers. In addition, as explained above, the LMD process requires a vertical nozzle in order to limit the effect of gravity on the powder jet and to keep the geometric characteristics of the deposited metal stable and reproducible. Apart from very simple geometry parts (made up of straight vertical walls), this constraint requires the implementation of 5-axis configurations (Figure 1.16) (Beam 2020):

- the nozzle moves on three linear axes, thus maintaining its vertical axis;

– the part moves on two rotary axes (named A and C or B and C) in order to orient the part under the beam and the molten material deposit and thus maintain a vertical stacking direction.

The choice of machine depends on the volume of the part to be produced, on its complexity and of course on the process implemented. The largest LMD machines on the market today are capable of handling parts over 1 m in length.

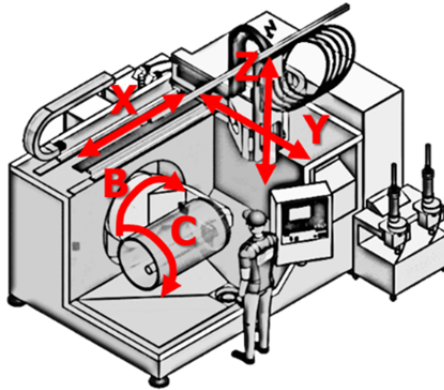


Figure 1.16. Example configuration of an industrial LMD machine (IREPA 2012).

For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

1.1.7.2. Hybrid machines

As with other additive manufacturing processes, several manufacturers have marketed machines that combine additive manufacturing and machining by removing material (DMG 2019). This machine concept is interesting because it makes it possible to be able to machine as the construction progresses; for example, surfaces that are no longer accessible or initial surfaces for a part built in a later phase. On the other hand, these machining operations cannot finish machining of the part for materials requiring heat treatments or parts requiring stress relieving treatment, because the geometry of the part is often impacted by the thermal cycle during processing. Other manufacturers have proposed hybrid concepts involving the deposition of powder and wire on the same machine (Meltio3D). This type of innovative and promising concept, still at a prototype stage, has yet to prove its value.

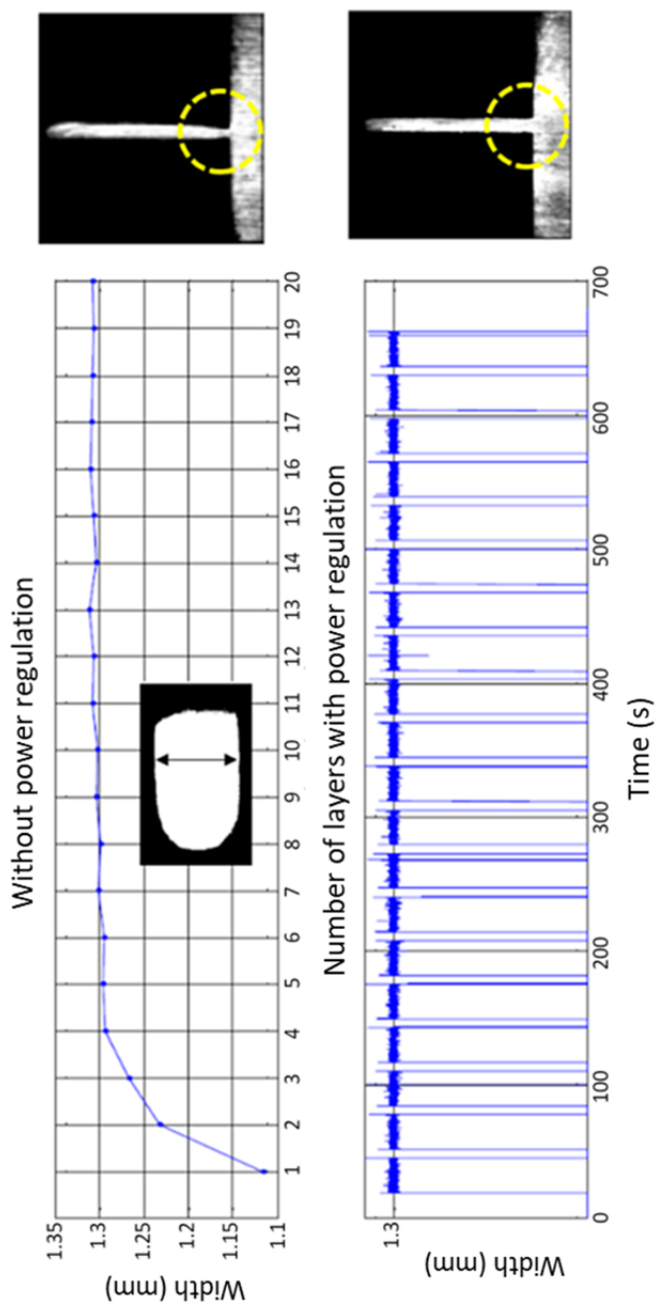


Figure 1.17. Control of the molten width by power regulation at the foot of a wall developed by LMD (Mezari 2014)

1.1.7.3. *Process monitoring and control*

Various types of real-time monitoring and control can be carried out during the process. This is the case for the monitoring of the powder jet via, for example, a diagnostic coupling a laser sheet and a camera attached to the projection head (LiSEC system from Fraunhofer IWS; Lisec 2019). This is also the case for the stability control of the FZ for which various coaxial observation systems are marketed. The sensors used can be photodiodes (average information) but they are most often C-Mos or CCD cameras (2D image), making it possible to extract the contour of the FZ and its characteristic dimensions (projected area, width, length, even deposited height). The challenge of such diagnostics is limiting the size drift of the FZ, for example, in turnaround zones (3D part edge) or “hairpin” trajectories favoring the accumulation of heat. The stabilization of the size of the FZ is most often done by regulating the laser power via closed-loop servo control (Mezari 2014), which leads to better controlled geometries (Figure 1.17).

1.1.7.4. *Industrial applications of the DED-LMD process*

The first industrial implementations of the LMD process dealt with the repair of aeronautical parts. For example, a French aircraft engine repairer invested in two machines to repair rotating Pratt and Whitney engine parts. With this type of part (rotating air/oil seal; Figure 1.18), it was a question of rebuilding the various lips, worn during their previous use. These lips are machined under the nominal dimension, reconstructed by material deposition (Ti alloy: Ti6242), heat treated and remanufactured before returning for normal service as a new part.



Figure 1.18. *Lip repair of an air/oil seal via LMD process*

One can find in the literature many implementation examples of the LMD process relating to prototype parts or parts produced in small series, in particular for

applications in the aeronautical and space sectors. For series applications, many developments are underway, in particular for the production of prototypes for parts close to the final dimensions. This type of application is very promising for small series because it allows parts to be manufactured quickly, limiting material loss and without having to manufacture tools. An application derived from the previous one consists of depositing material on a semi-finished part in order to provide it with new functionalities. Thus, starting from a part made by another process, it is possible to locally modify the geometry to strengthen it or to implant sensors therein.

Unlike powder bed processes, the process also allows easy manufacturing of gradient materials or composites from pre-alloyed powders or a mixture of mother powders. However, few applications are currently listed in this field.

Finally, other concentrated energy material deposition (DED) processes use lasers coupled to a wire deposition. These are WLAM processes. As for the WAAM processes, based on the fusion of a wire by electric arc (see section 1.4), they have the advantage of a material yield of 100% (all the molten wire contributes to the manufactured part) associated with the flexibility of laser processes. WLAM integrated heads are currently marketed by various manufacturers (Precitec 2019), with wire diameters between 0.5 and 2 mm.

1.2. The L-PBF process

1.2.1. Distinction between sintering and laser melting

Selective laser sintering (SLS) and selective laser melting (SLM or L-PBF) processes on metals, which use an identical laser beam, are differentiated by the consolidation mechanisms of the powder bed. These result from complete (SLM or L-PBF) or partial (SLS by liquid phase sintering) melting or complete non-melting (SLS by solid phase sintering). Sintering is a process that heats powder without leading to its complete melting, whereas L-PBF is a rapid melting or welding process. More information on SLS is available in section 1.5.

1.2.2. Manufacturer interests and requirements

The L-PBF process, used more and more by many industry sectors (aeronautics, space, biomedical, luxury, etc.), has reached such a level of maturity that certain aeronautical and space parts are now qualified for flight (Silvercrest engines from

Safran and Prometheus from ArianeGroup). Today, because of increasingly large manufacturing enclosures, manufacturers are seeking to produce small series of small-sized parts on the same platform. In addition, they are keen to obtain finished parts for which the dimensions, shape and surface conditions meet the pre-established specifications.

1.2.2.1. *Process advantages*

The L-PBF process has many advantages including freedom of design. Indeed, it is now possible to design parts with a much more complex geometry than those obtained by traditional shaping techniques (casting, forging, machining, etc.). Parts incorporating cavities, lattices and strong undercuts (Figure 1.19) can be made in a single operation, with the introduction of a support or decent compacting of the powder beds. Another advantage highly appreciated by manufacturers relates to the limitation of metallurgical assemblies of welded or brazed parts, which generate defects at the interfaces and weaken the assemblies.



Figure 1.19. *Example geometries of lightweight parts, suitable for layered production (Rias et al. 2014)*

In addition to the manufacture of single-piece parts, this additive process also makes it possible to produce convoluted channels, which are impossible to achieve with drilling or any other traditional technique. Thus, it improves the efficiency of heat removal from plastic or liquid metal injection molds by implanting cooling channels as close as possible to the molding surface of the mold cavity. This “conformal cooling” effect reduces the injection cycle time by 30% and improves the quality of the injected parts (Beard 2014; Marcos 2018). Finally, the absence of tools and the elimination of certain machining operations reduce costs and manufacturing times.

1.2.2.2. Disadvantages of the process

As the L-PBF process is a layer-by-layer fusion process, the manufactured part has a very complex thermal history, which generates significant residual stresses after cooling. These thermal stresses and strains are linked to differences in expansion between several areas of the construction where temperatures are different. Metallurgical and mechanical effects in the solid phase are often associated with this predominant thermal effect, with possible phase changes that can lead to significant volume variations, via transformation plasticity. However, L-PBF differs from other liquid forming processes (welding and foundry) in the thermal cycling experienced by the material and the significant remelting of the previous layers. It is also based on a triangular interaction scheme between thermal, metallurgical and mechanical phenomena (Figure 1.2).

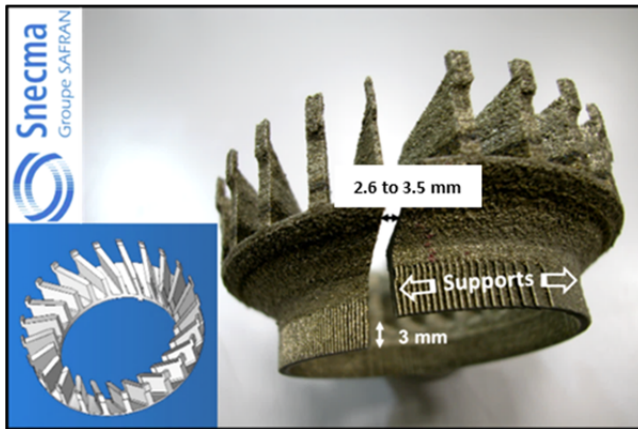


Figure 1.20. Deformation of an IN718 Safran injection spindle, made with L-PBF after wire cutting by spark erosion, revealing the presence of residual stresses

Figure 1.20 shows the distortions of an IN718 injection spindle made by L-PBF after wire cutting by spark erosion. Cutting the wire parallel to a fin causes a displacement of 3 mm and an opening in the twist in the cutting plane. This reflects the presence of internal shear stresses and ortho-radial tensile stresses. The operator then has two choices to reduce the residual stresses: either perform stress relief before separating the part from the build plate or optimize the construction strategy to reduce thermal gradients in each layer. Finally, another disadvantage comes from the anisotropy of the microstructure of the L-PBF prototypes in the direction of manufacture with a grain elongated in this direction. This “natural” orientation of

the grains can lead to anisotropy of the mechanical properties, not generally desired by manufacturers.

1.2.3. Process principle and basic elements

A typical machine consists of two tanks in which a piston moves to the nearest micrometer and can go up (supply piston) as well as down (production piston). The first of these two tanks is for feeding powder and the second is for manufacturing parts. Before any manufacture occurs, the feed tank, the piston of which is in its lowest position, is loaded with a previously steamed powder. As for the manufacturing tank, its piston is in its highest position (Figure 1.21). Other means of powder feeding, such as hoppers with or without slides, equip some industrial machines today. The powder circulates and then falls in front of the layering system.

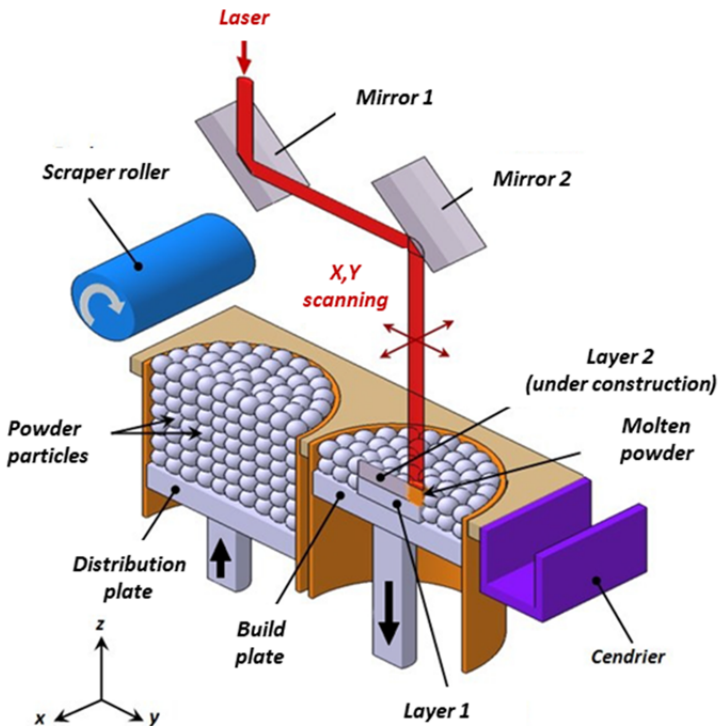


Figure 1.21. Principle of the L-PBF process (Orecchia 2008). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

1.2.3.1. Layering and consolidation cycles

Since L-PBF is a layer-by-layer manufacturing process, each new iteration corresponds to a layering cycle and a laser fusion consolidation cycle, which break down as follows (Figure 1.22).

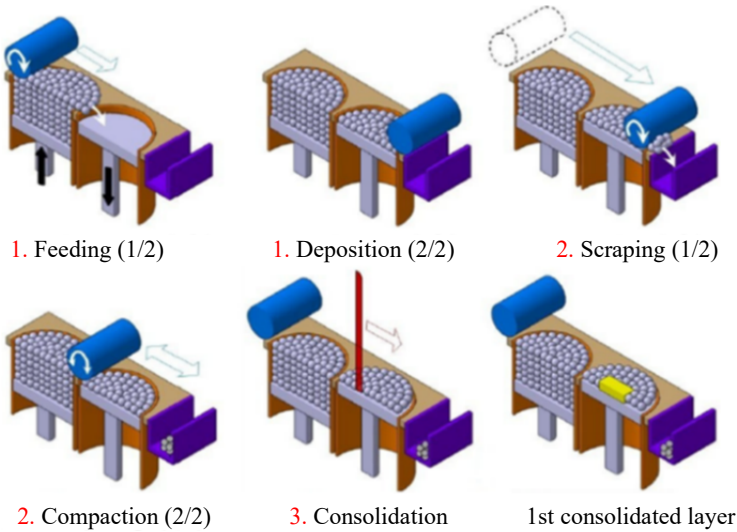


Figure 1.22. Main steps in the L-PBF manufacture of a part (Orecchia 2008).
For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

– Step 1 (feed and deposition): The feed piston rises to a certain height in order to supply enough powder to the level of the build plate and guarantee perfect spreading of the powder, thus obtaining a powder bed of homogeneous density and uniform thickness (ΔZ_{powder}). As for the build plate, it descends from a fixed height (ΔZ) corresponding to the slicing of the part and, after a few layers of stabilization, to the thickness of the powder consolidated by fusion. The finer the slicing, the better the surface finish of the parts and the finer the grain size of the powder.

– Step 2 (scraping and compacting): A scraper consisting of a roller, a carbon bristle brush or a rigid or flexible blade spreads and then levels the powder on the build plate. Some machines allow this step to vibrate or compact the powder with ultrasound (sonotrode) or using a roller compactor. All of this increases the compactness of the powder bed and therefore increases the average coordination number of a particle with its closest neighbors. The surplus powder is then poured into the ashtray. The latter can generally be reused after repackaging (sieving to eliminate any ejecta or “spatters” and particle agglomerates, etc.) (Vinson 2015).

– **Step 3 (consolidation by fusion):** A high-power laser beam scans the areas of the powder bed to be fused at each layer, in accordance with the “machine” parameters and the “manufacturing” parameters. The displacement of the laser spot is generally ensured by a scanner head made up of rotating mirrors and orientable through galvanometric motors, which make it possible to generate high scanning speeds V_0 in m/s.

– **Step 4 (new iteration):** When the layer has finished being printed, the layering system returns to its initial position and a new iteration starts until the complete production of the 3D digital model of the part.

1.2.3.2. *Focus on some basic elements*

– *Work atmosphere:* The manufacturing cell is constantly swept by a laminar flow of neutral gas (argon or nitrogen) at a slight overpressure during the manufacture of the part, without which surface oxidation of the melt-pool may occur. The permissible value of the partial pressure of oxygen P_{O_2} and of the water vapor P_{H_2O} is strongly dependent on the nature of the material to be fused and the desired metallurgical quality. Typically, so-called reactive alloys such as titanium alloys require special attention because oxygen dissolves easily in them. As for aluminum alloys, they are very sensitive to humidity because liquid aluminum, in contact with water vapor, forms gaseous hydrogen pores on solidification ($3 [H_2O]_g + 2 (Al)_l \rightarrow <Al_2O_3>_s + 3 [H_2]_g$) (Vilaro 2011). The neutral gas flow (preferably argon) has the dual interest of (1) inhibiting chemical reactions and (2) evacuating the fumes and dispersing the “spatters” generated by the vapors outside the work area, thus protecting the optics and limiting slag on the powder bed. The direction of flow of the shielding gas must then be perpendicular to the scanning direction of the laser beam and its flow rate must be optimized. For the same reason, the first parts to be printed should be farthest from the gas inlet (Ferrar et al. 2012). Some researchers (Masmoudi 2016) have replaced argon with a low vacuum (0.1 bar) and point to improved melt-pool stability, fewer spatters, less denudation (Matthews et al. 2016) and, indirectly, better material quality.

– *Build plate:* The build plate conditions the heat dissipation in the part, the thermal conductivity of the powder bed often being assumed negligible. Also, its nature, its thickness, its surface condition and its absorptivity can be important elements if the part is built directly on the plate without going through a support. This plate, fixed to the manufacturing piston, is often much more rigid and larger than the part. It therefore introduces significant internal stresses in the part and in particular at the plate/part interface. Its upper surface is most often sandblasted so that the powder particles cling to it more easily and its absorption is better. This surface should be flat and parallel to the movement of the layering system.

– *Preheating system*: To meet the growing demands of manufacturers (cracking materials), certain build plates have been fitted with a heating system (up to 650°C) in order to reduce the amplitude of thermal gradients and resulting residual stresses. However, excessive preheating can lead to sintering of the powder bed particles and oxidation problems. However, as this preheating system is not efficient enough for large parts, other systems have been developed. Thus, 3D Systems machines (formerly Phenix Systems) are equipped with radiators above the powder bed, ensuring preheating layer by layer.

– *Laser source and scan head*: The lasers used in L-PBF are mostly the single-mode fiber lasers Yb:YAG ($\lambda = 1.07 \mu\text{m}$) or Nd:YAG ($\lambda = 1.06 \mu\text{m}$), exhibiting Gaussian power distribution and beam diameters (at $1/e^2$) less than 100 μm . The laser describes, at speeds V_0 of m/s, the fusion trajectory thanks to a scan head comprised of two galvanometric mirrors, before it is focused thanks to an f-theta lens.

– *Multi-lasers*: In order to increase the productivity of the L-PBF process, manufacturers of industrial machines have continued to expand the manufacturing enclosures and since 2015 have equipped them with several laser heads. Today, it is common to work on machines with four lasers each delivering a power P_0 of 500 W.

– *Powder*: The powder strongly conditions the manufacture and metallurgical quality of a part. It is the subject of Chapter 2. It must have properties suitable for the L-PBF and the type of source used, have a good quality/price ratio and facilitate its reuse. The desired properties are accordance with the composition, inclusive cleanliness, fine and sufficiently large particle size, low intraparticle porosity, good sphericity and circularity (Vilaro 2011).

1.2.4. Types of materials involved

The range of materials manufactured by L-PBF is today quite wide, from alloys based on titanium, iron, cobalt, nickel, copper, aluminum or precious metals to certain refractory alloys with a high melting point. In general, the grades suitable for L-PBF manufacture are those said to be weldable, insensitive to conventional welding defects (hot and cold cracking). However, highly conductive materials (pure copper and aluminum), fragile materials with low toughness (intermetallics, ceramics) and so-called cracking materials (high carbon tool steels, nickel alloys highly loaded with hardening precipitates) are still difficult to control. Regarding highly conductive materials, it is their low absorptivity at $\lambda \sim 1 \mu\text{m}$ which is lacking. Either the wavelength must be adapted to the material (green lasers $\lambda \sim 0.51 \mu\text{m}$ or blue $\lambda \sim 0.45 \mu\text{m}$) or the power of the laser must be increased. Finally, concerning cracking materials, it is necessary to either limit the cooling rate by integrating an

efficient preheating system or to modify the composition of the alloy in order to make it less sensitive to hot cracking.

1.2.5. Presentation and influence of the process operating parameters

Obtaining good metallurgical quality parts that meet the desired dimensional and geometric precision requires perfect control of the process thermal cycles and therefore good adjustment of the operating parameters. In the case of the L-PBF process, these are classified into two categories: (1) the first-order parameters that are the most influential with regard to the stability of the melt-pool and in principle only concern the study of a single track, and (2) second-order parameters that are less easily changed during manufacture, but which also have a real effect on the quality of the printed part. They are often fixed at the start of manufacture according to the geometry of the part, the material considered and the desired metallurgical quality. The output parameters are used to judge the condition of a printed part by L-PBF, in particular from the geometric, dimensional, metallurgical and mechanical points of view.

1.2.5.1. Output parameters

First, the output parameters make it possible to assess the quality of the 1D bead, which constitutes the elementary block in any manufacture regardless of its geometric complexity, but also of a 2D wall or a 3D cube. This evaluation is carried out according to several criteria, namely the geometry of single beads that accounts for the build stability (see section 1.2.8), the metallurgical quality and the mechanical performance of the construction and finally according to an economic criterion. In this regard, it appears that the critical flow rate of construction D_v^{crit} ($\text{cm}^3 \cdot \text{h}^{-1}$) is the most meaningful for manufacturers. This economic criterion can then be approximated as follows:

$$D_v^{\text{crit}} = \Delta Z_{\text{plate}} \cdot D_0 \cdot V_0 \quad [1.2]$$

where ΔZ_{plate} is the incremental displacement of the build plate and D_0 the diameter and V_0 the scanning speed of the laser beam.

1.2.5.2. Input parameters and their effects on the bead dimensions

1.2.5.2.1. First-order parameters

– Laser power (P_0): The height H_{ZF} and width e_{app} of a melt-pool (MP) increase linearly with the laser power in the field of a stable MP. Indeed, a greater quantity of powder is melted when P_0 increases. Thus, the melt pool has a larger volume, but

not necessarily a higher average temperature, which has the effect of reducing the thermal gradients in the MP.

- Scan velocity (V_0): Unlike power, the height H_{ZF} and width e_{app} of the melt-pool decrease sharply as the speed increases. This is easily explained by the decrease in interaction time ($t_{int} = D_0/V_0$) between the laser beam and the material, which induces a lower energy input and therefore a smaller and cooler melt-pool. This then results in a higher cooling rate of the MP.

- Diameter of the laser beam (D_0): It is rather the energy distribution of the laser beam than its diameter that affects the morphology, width and thermal cycle of laser fusion beads. On average, the widths of the beads e_{app} vary between $1.25 D_0$ and $3 D_0$, the latter, and therefore the associated fusion isotherm, widening with linear energy or energy density (Vilaro 2011).

- Moving of the build plate (ΔZ): We consider here ΔZ or ΔZ_{powder}^1 ; these sizes have a direct effect on the apparent height H_{app} and the depth H_{ZR} of the beads, but also the number of melted layers $N_{CR} = H_{ZR}/H_{app}$. For increasing layer heights and at constant volume energy (E_v), the number of molten layers then logically decreases, the molten depths being linearly dependent on E_v (Fabbro 2019).

1.2.5.2.2. Second-order parameters

- Build method: The method is subdivided into two stages, namely (i) the scanning strategy within a layer that corresponds to the path described by the laser beam on a surface of a given layer and (ii) the layer stacking method involving the layering of beads, layer by layer.

- Scanning strategy: This strategy involves the hatch parameter e_v , which represents the distance between two adjacent laser tracks (or beads) of the same layer. The overlap ratio τ_R combines the apparent bead width and the hatch across the relation:

$$\tau_R = 1 - \left(\frac{e_v}{e_{app}} \right) \quad [1.3]$$

This overlap ratio is a function of the linear energy or energy density, which indicates the percentage of remelted material in the adjacent bead. The closer e_v is to e_{app} , the smaller the τ_R and the shorter the manufacture time.

1. As a reminder, due to shrinkage on solidification linked to the compactness of the powder bed (about 50%), it is often considered that $\Delta Z_{powder} = 2 \Delta Z$ (axial shrinkage only). Thus, a plate increment of 50 μm corresponds to 100 μm of spread powder.

– Stacking method: The $0^\circ/90^\circ$ cross method is certainly the most widely used stacking strategy and is similar to the multi-angle strategy (the angle varies with each layer; see Figure 1.23) because it limits the appearance of defects (lack of material) and decreases the construction anisotropy. Another less widespread strategy known as the “staggered on two layers” method (Lefeuvre 2019) is possible. It corresponds to a stack of scanned layers with the same $0^\circ/0^\circ$ orientation. The vectors are not superimposed on each other but shifted by $1/2 e_v$ between each layer so that a bead of the N layer comes between two beads of the previous N – 1 layer (Figure 1.24). This method thus makes it possible to overcome the lack of powder due to denudation.

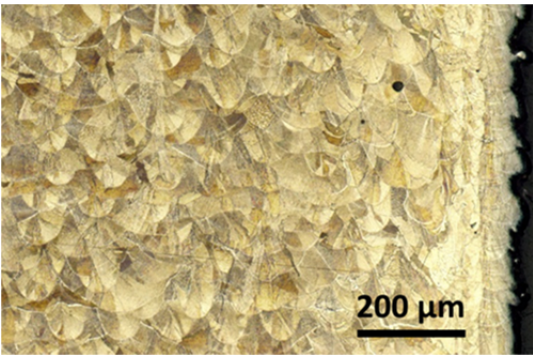


Figure 1.23. Example of L-PBF microstructure on 316L stainless steel at the scale of the beads (multi-angle strategy, distinction between contour and part core)

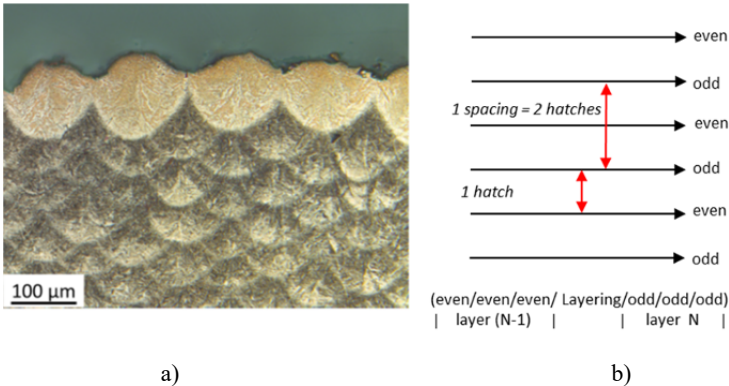


Figure 1.24. a) Optical image of bead stacking ($\tau_R = 33 \%$);
b) associated “staggered on two layers” method (Lefeuvre 2019)

1.2.5.3. Combined parameters

The combination of several of these parameters (P_0 , V_0 , D_0 , Δz and e_v) gives rise to energy parameters that correlate rather well with construction stability, bead shape and dimensions, porosity rate, crack density, etc. These combined parameters, which do not always correspond to a physical reality, are written as follows (with t_{int} = interaction time = D_0/V_0):

- power density ($\text{W} \cdot \text{mm}^{-2}$):

$$I_0 = 4 \cdot P_0 / (\pi \cdot D_0^2) \quad [1.4]$$

- linear energy ($\text{J} \cdot \text{mm}^{-1}$):

$$E_l = P_0 / V_0 \quad [1.5]$$

- surface energy ($\text{J} \cdot \text{mm}^{-2}$):

$$E_s = I_0 \cdot t_{\text{int}} = 4 \cdot P_0 / (\pi \cdot D_0 \cdot V_0) \quad [1.6]$$

- volume energy density ($\text{J} \cdot \text{mm}^{-3}$):

$$E_v = I_0 / V_0 = 4 \cdot P_0 / (\pi \cdot D_0^2 \cdot V_0) \quad [1.7]$$

- overall energy per volume produced ($\text{J} \cdot \text{mm}^{-3}$):

$$E_v^g = P_0 / (e_v \cdot \Delta z \cdot V_0) \quad [1.8]$$

The first volume energy E_v reflects the intensity of the laser fusion (it linearly conditions the melted depth) of the single bead, and the second E_v^g corresponds to energy injected per volume of material manufactured.

1.2.6. Comparison of the L-PBF process and DED process

Table 1.2 compares the energy parameters as well as the thermal and dimensional variables of a melt-pool or a 1D bead (elementary block in the manufacture of a part) for two methods. The main difference between these two methods is the value of the operating parameters (P_0 , V_0 , D_m , D_0) and the average diameter (d_{50}) of the powder particles. Both the spot size and the laser power are much weaker in L-PBF than in DED. The scan velocity is much greater in L-PBF than in DED because of the use of galvanometric mirrors. Finally, the average diameter of the powders is about five times smaller in L-PBF than in DED. These differences result in different energy inputs and different laser/material interaction times and therefore ultimately different thermal quantities (G , V_s and $\dot{T}$), stress states and microstructures specific to each process.

	DED	L-PBF
Laser power, P_0 (W)	200–2,000	20–200
Scan velocity, V_0	2.5–25	100–1,500
Laser beam diameter, D_0 (μm)	1,500	50
Average powder diameter, d_{50} (μm)	25–125	5–25
Bead height, H_{app} (μm)	500	50
Bead width, e_{app} (μm)	2,500	250
Interaction time ($= D_0/V_0$), t_{int} (ms)	60–600	0.03–0.3
Laser power density, I_0	$\approx 10^2$ – 10^3	$\approx 10^4$ – 10^5
Linear energy ($= P_0/V_0$), E_L ($\text{J}\cdot\text{mm}^{-1}$)	8–800	0.04–4
Energy density, $E_V = I_0/V_0$ ($\text{J}\cdot\text{mm}^{-3}$)	≈ 4 –400	≈ 20 –2,000
Solidification rate, V_s ($\text{cm}\cdot\text{s}^{-1}$)	$\approx 10^{-1}$ – 10^0	$\approx 10^2$ – 10^3
Cooling rate, $\dot{T}$ ($\text{K}\cdot\text{s}^{-1}$)	$\approx 10^2$ – 10^4	$\approx 10^5$ – 10^8

Table 1.2. Comparison of DED and L-PBF processes

1.2.7. Design and manufacturing method of a part

1.2.7.1. Shell, core and contour

When manufacturing a part made up of thin and solid parts, a specific scanning strategy for the solid parts is often used that consists of assigning them a shell and a core (Figure 1.25). Overall, optimizing the manufacturing of the core as well as the contour aims to reduce the pore level and avoid hot spots, which cause deformation or residual stresses. Generally, the shell, comprising several contours, is made with a set of operating parameters that are less energetic than the core. On the outer surface of the shell, a final contour (upskin) is sometimes produced with an adapted set of parameters, the objective of which is to guarantee the dimensions of the part and reduce its roughness. There are a multitude of scan strategies (Figure 1.26) for filling the core of the part. The most commonly used are striped or chess strategies. The width of the bands as well as the size of the boxes in the checkered pattern is around a few millimeters, but remains a variable for each of these two strategies. The orientation of bands and vectors within these bands is also a variable that has a

marked effect on the texture of the material. You can also choose to alternate the scanning directions, or even modify the order of lasing (Figure 1.26(g)). The lasing time will then be longer or shorter, depending on the configuration chosen.

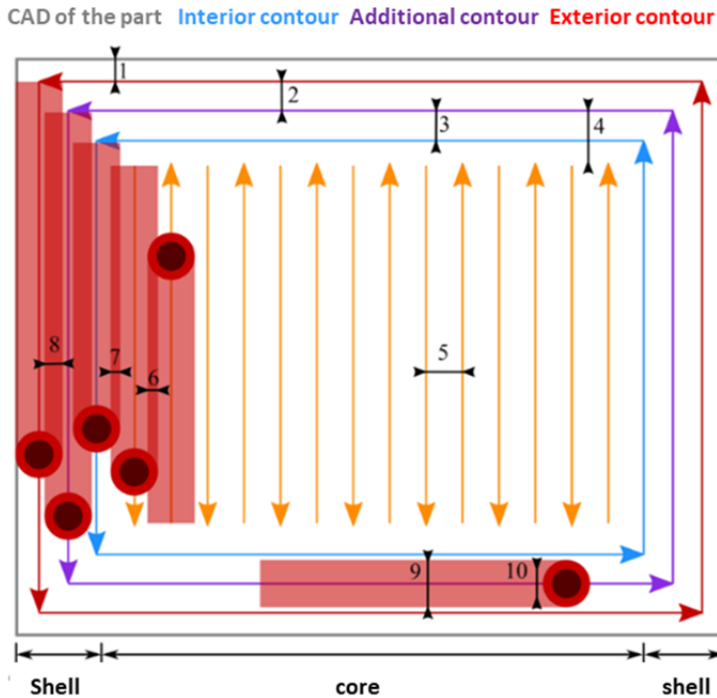


Figure 1.25. Example of a core/shell manufacturing method: (1) laser beam compensation; (2–4) contour, inner contour and fill-wise zone offsets; (5) hatch (ev); (6–8) bead/bead, bead/contour and contour/contour overlaps; (9) effective laser diameter (bead width); (10) laser diameter. For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

The melting order of the different part components can also be modified directly by the operator (contour/fill or fill/contour). Likewise, the melting order on the plate is often set up toward the incident gas flow (the most distant parts first) in order to limit the interaction between the laser and the metal fumes. Overall, poor management of all of these parameters can greatly affect the number of pores and their location, with equivalent operating parameters. Thus, the multiplication of transition zones generates a large number of defects while the random scanning of the chess pattern allows more homogeneous heating and less internal stresses. For

large surfaces, the discretization in millimeter bands makes it possible to limit long beads, which generate instabilities and internal stresses. However, for $0^\circ/90^\circ$ layer stacks, this strategy is unsuitable because pores are generated at the bands' intersection. This type of defect is usually mitigated for rotations with a “prime” angle of 67° at each layer (Figure 1.27).

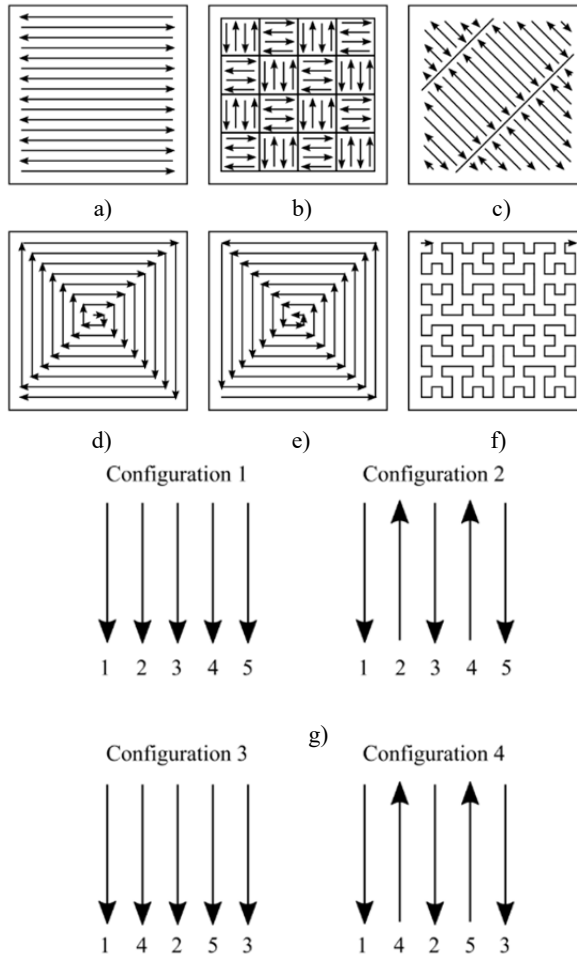


Figure 1.26. Types of methods for filling massive parts: a) back and forth, b) checkered or islands, c) bands, d) outward spiral, e) inward spiral, f) Hilbert fractal and g) four scanning strategies (variation in direction and order). Configurations 2 and 3: minimum and maximum laser time

Another frequently encountered issue is related to the stability of the molten pool of each laser vector associated with the inertia of the scanner head. In the old “delay” fusion mode, the acceleration–deceleration phases upstream and downstream of the laser line (reduced scan velocity) contributed to increasing the linear energy and therefore the dimensions of the melt-pools. One solution is the “sky-writing” method for which the path of the scanner head is initiated outside the part (before and after lasing), without the laser being switched on, in order to ensure a constant scanning speed.

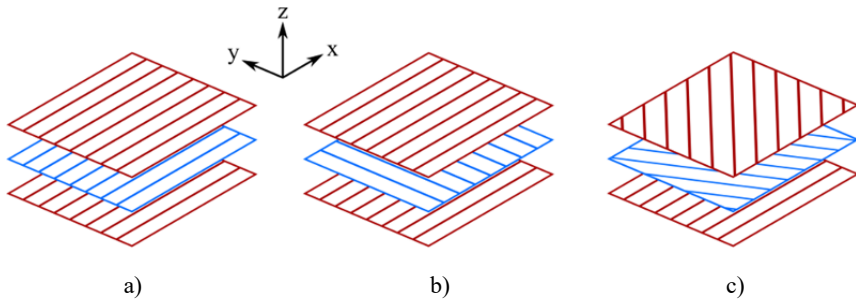


Figure 1.27. Layer stacking strategies: a) 0° at each layer, b) alternating orientations of 90° at each layer and c) 67° at each layer. For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

1.2.7.2. Manufacturing supports

Manufacturing supports are one of the constraints of the L-PBF process that must be incorporated from the design and positioning of the part on the build plate. These are friable elements that are easily destructible using pliers and produced at the same time as the part, necessarily with a less energetic parameter than those used for the manufacture of the part. The supports are placed below surfaces that are “downskin” to the build plate surface (Figure 1.28). The support limit angle, close to 45° , depends on the material and the manufacturing parameters. The supports must avoid the deformation or even the collapse of the “downskin” surfaces in contact with the powder, linked to the force induced by the passage of the layering system and the weight of the solid parts. For these massive areas, the supports also serve as heat dissipators, with the powder assumed to be an insulating medium. Another role of the supports is to stiffen the thin parts, which are subject to deformation during construction.

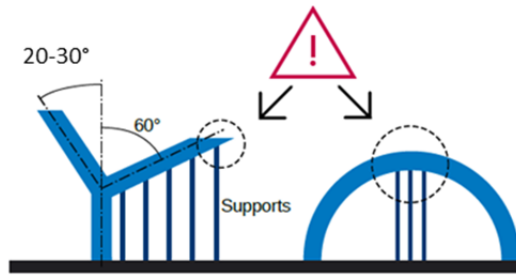


Figure 1.28. Example of geometries requiring supports (Marcos 2018)

1.2.7.3. Design rules

The construction of a complex shaped part via the L-PBF process requires compliance with certain design rules with regard to its geometry, its positioning on the production platform and the post-lasing finishing steps.

- Massive parts should be limited: They are often the site of defects (pores), due to the many hot spots and transition zones that are difficult to manage during the fusion. It is better to make thinner pieces with openwork shapes in order to limit the support.

- The hollow volumes inside a part (cooling channels, etc.) must exceed a certain diameter and have a moderate tortuosity, in order to facilitate depowdering and limit sintering of the powder. A powder outlet is then appropriate.

- The virtual positioning of the part on the production platform is done by minimizing the construction height and the number of undercut surfaces (beyond 40–45° inclination). This helps to reduce manufacturing time, powder consumption and any additional rework costs.

- It is advisable to keep a minimum and identical distance between fused parts in order to avoid ejecta and thermal interactions. Otherwise, internal stresses and the defect rate may vary depending on the part.

- The parts must be inclined at a certain angle with respect to the layering axis, in order to avoid damaging them by minimizing the contact and friction surface between the part and the scraper or roller during spreading.

- To manage the removal of supports or rework, it is necessary to promote accessibility to the part early at the design stage.

described by a parabola arc, which reflects a strong dilution. Such MP depths remain suitable for manufacturing and ensure good “material health” for the part. However, it is advisable to avoid the deep keyhole domain, which appears when the linear energy is too high ($> E_1^{\text{keyhole}}$), and the instability of which favors the formation of occluded cavities. Another hydrodynamic instability occurs at high power/high speed (Figure 1.29), what is known as “humping”.

Its critical speed of onset can be approximated by the following relation (Seiler et al. 2016):

$$V_{crit}^{hump} \sim \left(\frac{L_{ZF}}{e_{ZF}} \right) \cdot \sqrt{\frac{\gamma_{LG}}{e_{ZF} \cdot \rho_l}} \quad [1.9]$$

where L_{ZF} is the length of liquid disturbed by the flow, e_{ZF} is the melt-pool width, ρ_l is the density of the liquid and γ_{LG} is the liquid–gas interface voltage.

1.2.9. Optimization of L-PBF manufacturing of 3D parts

Once the manufacturing of stable single beads has been optimized, the most complex step remains: optimization of the manufacturing method in order to obtain the densest 3D part possible and/or a good compromise between manufacturing speed and density. The initial step involves optimizing the hatch e_v from the bead width by considering, for example, $e_v \sim 0.75 e_{ZF}$ (i.e. a recovery rate of 25%), which is practiced in many optimization ranges. The same type of approach, aimed at optimizing inter-bead overlap, was proposed analytically by Tang et al. (2017). Compared to the single-bead optimization (green part of Figure 1.29), it is then necessary to adjust the linear energy P/V in order to densify the parts while limiting the number of liquid ejecta that contaminate the powder bed and promote pores. The choice of the filling strategy (stripes, chess pattern) depends on the required outcome; a chess pattern can help reduce the thermal deformations and the stripe pattern the porosity rates, at equivalent P/V .

An example of optimization of the porosity rate on a 316L steel produced by L-PBF is presented in Figure 1.30, for three different layer heights Δh (or plate increments ΔZ_{plate}) (30, 60 and 100 μm). In all cases, densification begins beyond a volume energy density (VED) threshold, then follows an evolution in $1/\text{VED}^2$ to reach a minimum porosity rate of between 0.05% (layer height of 30 μm) and 0.2% (layers of 100 μm). This type of development reflects increasingly pronounced

inter-bead overlaps (because the size of the L-PBF single-beads increases linearly with VED) and a gradual limitation of lack of fusion.

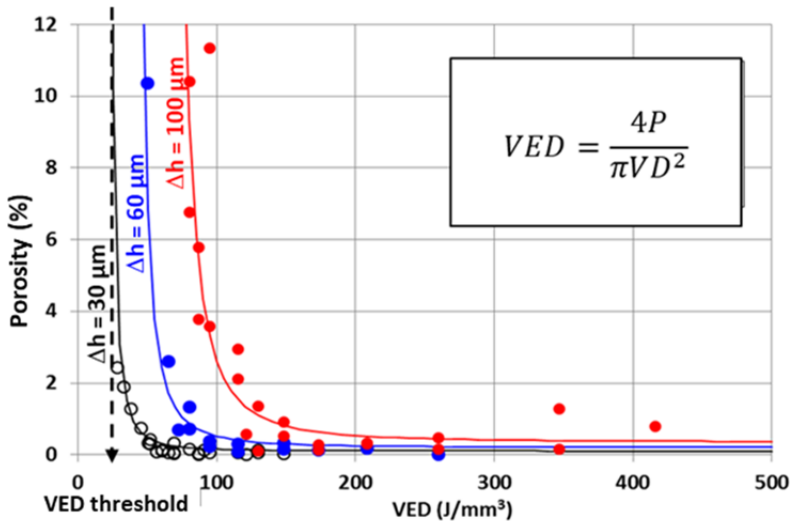


Figure 1.30. Porosity level as a function of volume energy density (equation [1.7]) and layer height Δh (Gunenthiram 2018). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

1.3. Electron powder bed fusion

1.3.1. Introduction

Electron powder bed fusion (E-PBF) is a powder bed metal additive manufacturing technique first marketed under the name electron beam melting (EBM) by the Swedish company Arcam AB (Arcam 2020) and since then acquired by General Electric (GE). Until very recently, Arcam was the only stakeholder in the market offering E-PBF machines. Today, other companies (Q-beam3D (China) and JEOL (Japan)) are developing the technology. Nevertheless, all the results presented in this chapter were obtained with Arcam AB machines, which represent more than 90% of the E-PBF machines. The E-PBF process allows the manufacture of components with excellent density (>99.9%). The technology looks a lot like the more popular L-PBF technology, but it uses an electron beam (EB) as the energy source. As in L-PBF, a layer of powder is deposited on a metallic substrate, but

locally melted with a moving electron beam. However, the use of an EB introduces some specificities. Thus, an EB can be deflected at speeds much higher than those used in laser processes, up to $10^5 \text{ m}\cdot\text{s}^{-1}$ because of electro-magnetic lenses. On the other hand, the process imposes a fairly high vacuum and can present instabilities because of electrostatic repulsion phenomena between the powder particles. In addition, the E-PBF process is exclusively reserved for conductive materials. Its main applications concern, like most other additive manufacturing processes, on the one hand, the biomedical field (custom-made titanium prostheses), and, on the other hand, the aeronautical field for components whose buy-to-fly ratio is critical. The objective of this section is to describe the E-PBF process while identifying its main characteristics.

1.3.2. Implementation of the E-PBF process

1.3.2.1. Preamble

As with other AM processes, the part or component to be manufactured is first designed with a computer (1) then discretized into layers (2) before being manufactured layer by layer in an E-PBF machine (3) and finally powdered at the end of the manufacturing process (4), in a process quite similar to that used in the L-PBF process. The parts are built layer by layer following the four steps shown in Figure 1.31. The main difference with L-PBF is the presence of a preheating phase with the defocused electron beam (Figure 1.31), which we will come back to later (Fabbro 2019).

Once the larger part is finished, the build chamber is cooled and the part is depowdered. The hourly manufacturing rates for the E-PBF process are between $70 \text{ cm}^3\cdot\text{h}^{-1}$ for first generation machines and $100 \text{ cm}^3\cdot\text{h}^{-1}$ for subsequent generations. Another difference from L-PBF is that the E-PBF process takes place under a secondary vacuum (10^{-4} – 10^{-5} mbar in minimum value), a direct consequence of the use of an electron beam as an energy source. Maintaining a vacuum during manufacture is an advantage for metallic materials reactive with O, N or even H. In certain cases (titanium alloys), helium is introduced into the chamber to maintain a constant pressure of 2×10^{-3} mbar in order to limit the electrostatic phenomena, which destabilize the process (see Chapter 3 of this volume) (Cordero et al. 2017).

Figure 1.32 is a photograph of the inside of an Arcam E-PBF machine chamber (model A1).

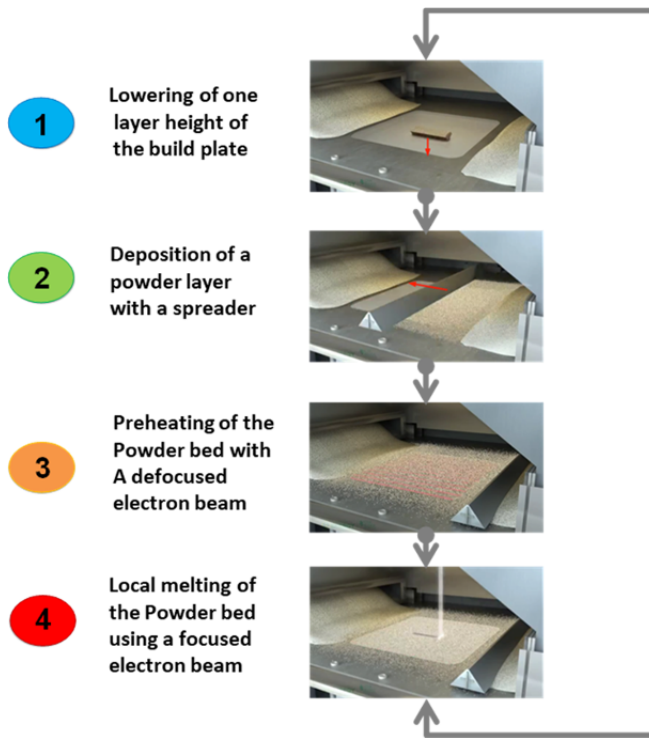


Figure 1.31. Manufacture using the four-step E-PBF process: (1) lowering of the layer height on the build plate; (2) deposition of a layer; (3) preheating of the powder bed by the defocused EB; (4) local melting of the powder bed with a focused EB. For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

The energy source comes from an electron gun similar to those used in electron microscopy, but using currents 10–100 times higher (1–50 mA). Electrons are emitted by a cathode made of a tungsten filament for early versions of E-PBF machines (Arcam A1, A2) and LaB_6 or CeB_6 for more recent machines (Arcam Q10, Q20, SpectraH). The electrons are then subjected to an accelerating voltage of 60 kV to generate a high-energy electron beam. The beam then passes through different electromagnetic coils in order to (i) correct the aberrations (astigmatism), (ii) deflect the beam to generate the trajectories and (iii) obtain a focused beam on the build plate (Figure 1.32(a)). The powers generated ($U \times I$) can reach up to 6,000 W for the latest generation of machines. While tungsten cathodes have the disadvantage of having a variable beam size with power (300–400 μm at 1 mm at high powers), LaB_6 cathodes offer a fixed beam size of 150–200 μm . The minimum dimensions that can be manufactured by this process are then around 500–600 μm .

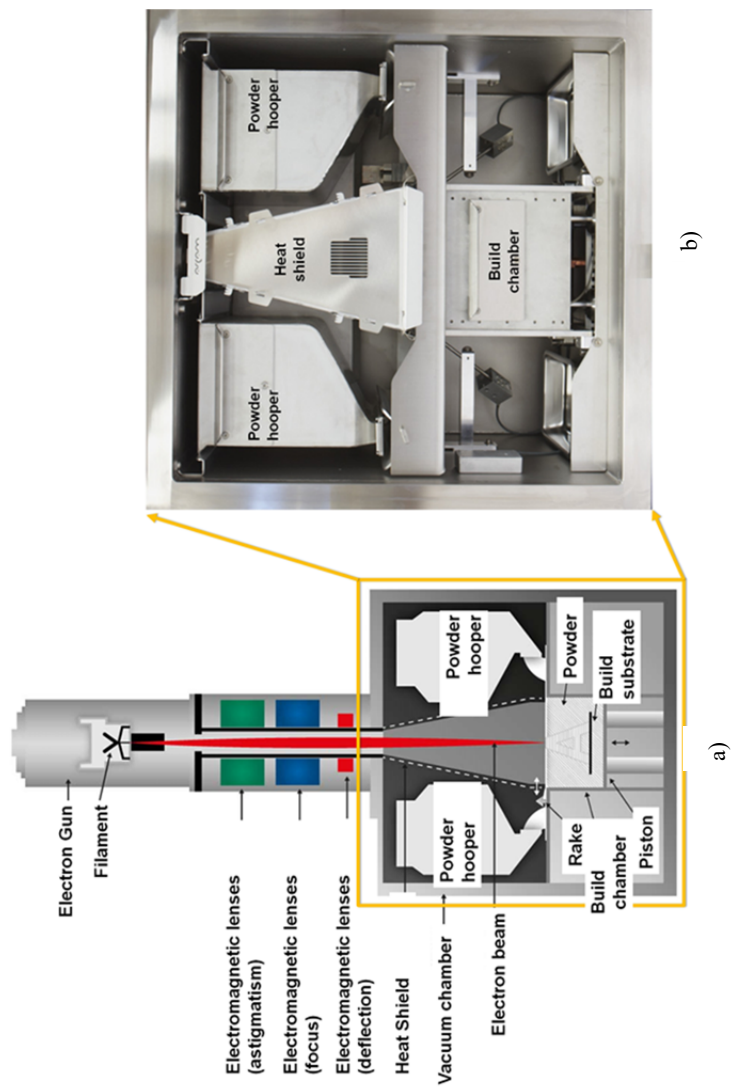


Figure 1.32. a) Representation of an E-PBF machine showing the essential elements. b) Inside of an E-PBF machine at Arcam

1.3.2.2. *Powders and layering*

The first step in the process is layering. The build plate descends according to the chosen layer height. The powder bed, of variable and controlled thickness, typically between 50 and 200 μm , is deposited using a rake which spreads the powder stored in the powder reservoirs located in the upper part of the build chamber.

To achieve a stable process, the powders used in E-PBF must have a spherical morphology and a typical size distribution of 40–100 μm resulting in homogeneous spreading of the powder bed. Small diameters should be avoided because they are more sensitive to “smoke” phenomena (Figure 1.33). Larger sizes are possible, but will result in poor surface finishes. Often the nature of the build plate is identical to that of the material being manufactured. However, we can deliberately choose a material different to the powder, for example, to facilitate the removal of manufactured parts.

1.3.2.3. *Preheating step*

The melting of the first layer is preceded by preheating of the build plate via the EB. Any new layer deposited will also undergo periods of preheating. These crucial steps are one of the specifics of the E-PBF process. Their objective is twofold: (1) to maintain the same average temperature at each layer throughout the manufacturing process, and (2) to achieve a slight consolidation of the powder bed. This consolidation makes it possible to increase the electrical conductivity of the powder bed in order to evacuate the electrical charges and avoid “smoke” phenomena (Cordero et al. 2017). This undesirable phenomenon can be summarized as follows: (1) during EB melting, a flow of negative charges is concentrated locally on the powder particles; (2) if the interparticle electrical conduction is insufficient, they will charge negatively; (3) when the repulsive electrostatic interactions become greater than the gravitational forces (mass of the powder particles) and the Van der Waals forces between particles, there is the formation of a powder cloud or “smoke” phenomena, very problematic because it generally leads to a sudden stop of the process (Figure 1.33). In summary, preheating is both essential and complex because it is necessary to find the right compromise between a sufficiently high temperature, which allows consolidation of the powder bed to avoid “smoke” phenomena in a reasonable time, and a sufficiently low temperature, avoiding the densification of the powder bed, thus making the depowdering steps tedious or even impossible. During these preheating steps, a defocused electron beam sweeps across the surface of the powder bed several times at high speed (typically between 10 and 60 $\text{m}\cdot\text{s}^{-1}$) and at high power (3 kW) (Figure 1.34(a)). The preheating temperature as well as the preheating methods (trajectories followed by the EB) depend on the material and the characteristics of the powders: density, size, morphology and electrical and thermal conductivity. Thus, in the case of a large fraction of small particles, resulting from a

low-density material with a low electrical conductivity (titanium alloys), the preheating must be adapted in order to avoid the formation of “smoke” phenomena. The construction temperature therefore varies significantly from one material to another: around 500°C for copper (Lodes et al. 2015) and up to 1,000°C for some Ni-based superalloys (Helmer et al. 2016; Chauvet et al. 2018a), or even 1,100°C for TiAl type intermetallics (Juechter et al. 2018).

1.3.2.4. Selective melting

Once the layer is preheated, the EB follows the trajectories defined in the manufacturing file in order to selectively melt the surfaces corresponding to the parts to be manufactured. Unlike the preheating step, the beam is focused and travels at slower speeds, from 1 to 8 m·s⁻¹. The most typical melting process is to first blend the contours (Figure 1.34(b)), then the interior surfaces (Figure 1.34(c)). The contours are often blended using the “multi-beam” mode, allowing the formation of several dozen simultaneous FZs (Figure 1.34(b)). It should be noted here that this is actually only one beam, but deflected so quickly that several FZs can be generated nearly at the same time. The interior surfaces are melted by a more classic “hatching” filling strategy (Figure 1.33(c)) with a direction that can change with each layer.

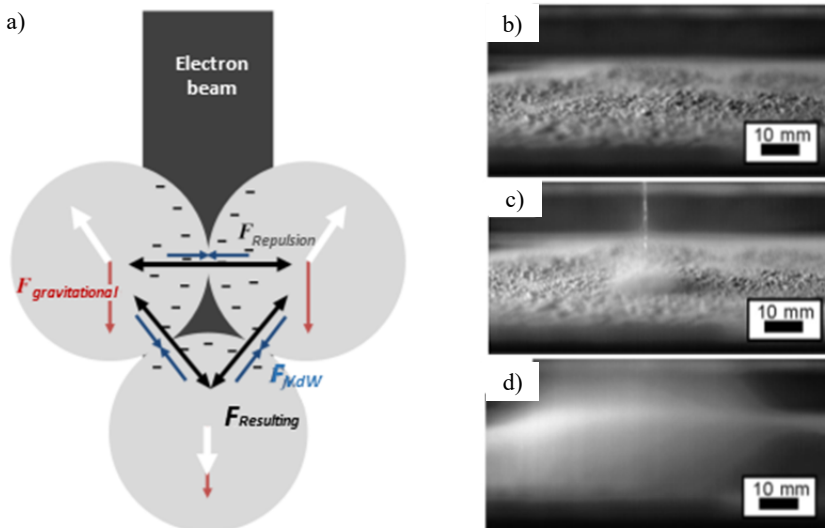


Figure 1.33. a) “Smoke” phenomenon observed using a high-speed camera during the E-PBF process: due to the electrostatic repulsion forces between negatively charged powder particles, a powder cloud is formed. b) Powder bed before EB exposure; c) irradiation by EB; d) formation of “smoke” phenomenon. From Sames et al. (2016). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

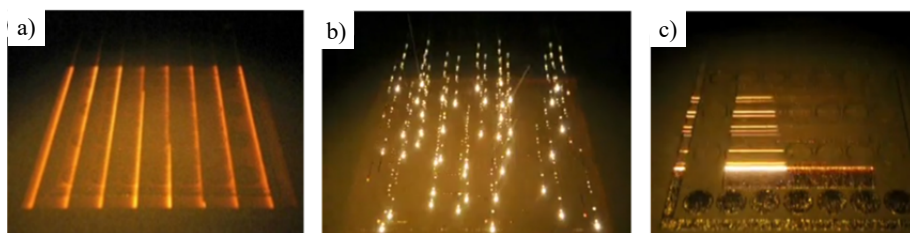


Figure 1.34. Fusion methods used at different key points in the E-PBF process. a) Preheating the powder with a defocused beam which sweeps the bed of powders several times in order to achieve consolidation; b) contour fusion using multi-beam technology; c) filling of surfaces by hatching

During the selective fusion step, many parameters can be defined: beam power P_0 ($= U \times I$, with U = acceleration voltage (V) and I = EB current (A)), beam focus, scan velocity V_0 , distance between fusion beads e_v , number of contours, bead melting order, etc. The same type of energy variables (E_v ($\text{J} \cdot \text{m}^{-1}$), E_s ($\text{J} \cdot \text{m}^{-2}$), E_v ($\text{J} \cdot \text{m}^{-3}$) as for L-PBF are proposed for parametric sensitivity studies (section 1.2.5.3).

As in L-PBF (Figures 1.25–1.27), the typical strategy used involves first melting the contours, then filling the areas to be melted by zig-zag hatching; the direction of this hatching changes, usually 90° with each layer. To date, little work has focused on original melting strategies, but some spectacular results have already demonstrated the importance of the melting strategy on the microstructures produced, especially for nickel-based superalloys (Helmer et al. 2016; Raghavan et al. 2017).

1.3.2.5. A few words on electron–matter interactions

The E-PBF process, like the L-PBF process, can be regarded as a microwelding process that allows the manufacture of parts made up of several thousand or tens of thousands of weld beads per cm^3 .

The major difference between laser and EB fusion comes from the absorption of the beam, which occurs on greater material thicknesses in E-PBF ($\sim 10\text{--}50\ \mu\text{m}$ vs. a few nanometers in L-PBF; see Chapter 3 of this volume and Klassen et al. (2014) for more details on electron–matter interactions). This results in less vaporization in E-PBF, which happens via conduction welding in E-PBF but via keyhole welding in L-PBF.

1.3.2.6. In situ control

The industrialization of additive manufacturing processes is currently hampered by process control and repeatability. Thus, to ensure the final quality of the parts, many

parameters must be monitored during manufacture. In E-PBF, *in situ* monitoring devices can be used, such as optical cameras and infrared thermography (Raplee et al. 2017), which provide information on the thermal cycles of parts during manufacture (Figure 1.35(a)). A remarkable specificity of the control of the E-PBF process is the use of the backscattered electrons to image the layer under construction, as with a scanning electron microscope (Pobel et al. 2019) (Figure 1.35(b)).

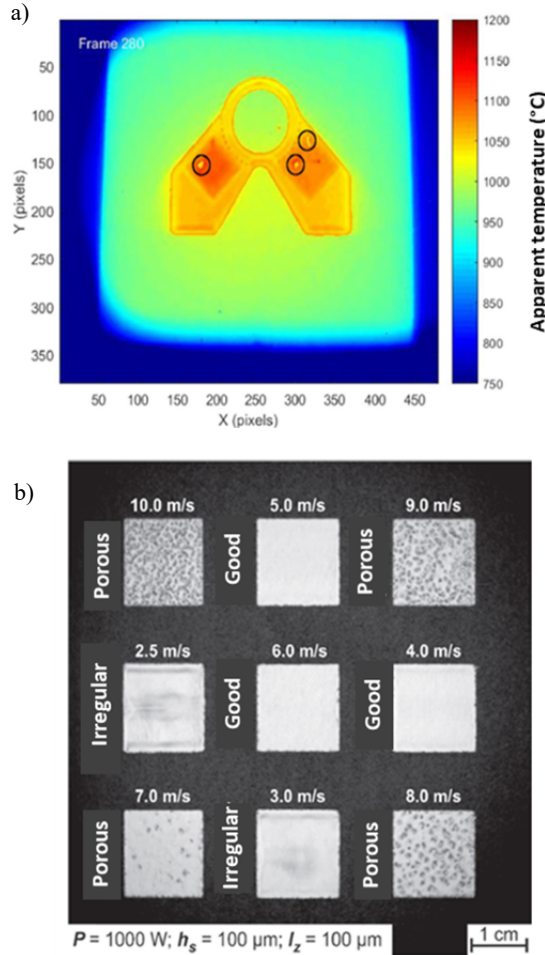


Figure 1.35. Examples of *in situ* monitoring a) by thermal camera, adapted from Raplee et al. (2017), and b) by electronic imaging (backscattered electrons), according to Pobel et al. (2019). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

1.3.2.7. Depowdering and recycling

As pointed out previously, one of the peculiarities of the E-PBF process is the consolidation of the powder bed by preheating before local melting to avoid “smoke” phenomena. The preheating temperature must be adjusted to allow consolidation of the powder bed (creation of necks between certain powder particles) while avoiding densification (growth of necks). The consolidation kinetics of the powder bed depend on the chemical composition of the powder, its physical properties and its surface properties, in particular any oxides which form on the surface. As a result of this preheating step, the manufactured parts are trapped in a consolidated powder bed (a “powder cake”) (Figure 1.36), which then requires an additional depowdering step by sandblasting with the same powder as that used during manufacture.

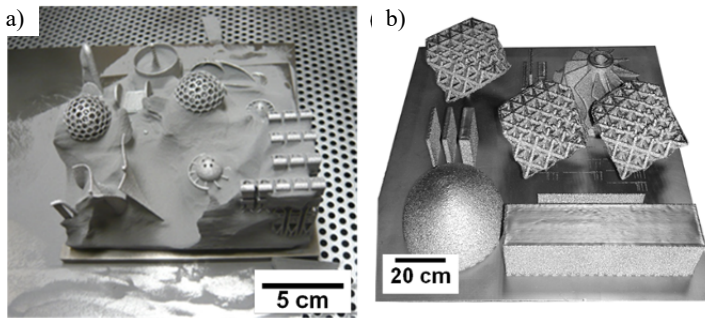


Figure 1.36. Photographs illustrating a) a “powder cake” during the powdering step and b) a build plate after powdering

This depowdering is delicate, in particular for tortuous internal channels or complex shapes. The recycling of the majority of the powder for its use in subsequent productions involves sieving, which can often break the necks between particles. Its characteristics (morphology, chemical composition) can then change during recycling (Tang et al. 2015; Gruber et al. 2019). This consolidation of the powder bed during preheating also has the advantage of limiting the use of supports (tedious to remove) because the consolidated powder bed has a certain mechanical strength. It also helps dissipate heat.

1.3.3. Optimization of melting conditions and characteristic defects

1.3.3.1. Optimization of melting conditions

In order to obtain dense materials, the FZ must be much deeper than the layer thickness (Bauereiß et al. 2014). Experimentally, a depth of three to four times the

layer thickness thus avoids the presence of porosities between layers. This *a priori* high number is explained by the thickness inhomogeneity of the powder bed before melting and highlights the importance of the layering device. Optimization of fusion parameters (i.e. power and speed) often involves design of experiments in which the energy parameters (E_l , E_s and E_v) span a wide energy range. Depending on the energy input (Figure 1.37), the samples are often classified into three categories: (1) porous samples with an “orange peel” surface with densities <98–99% and associated insufficient energy supply leading to “lack of fusion” type defects (Figure 1.38(a)); (2) dense samples (>99.5%, without lack of fusion), presenting a flat upper surface, associated with reasonable energy input; (3) samples with a high density (>99.5%, Figure 1.38(b)) but a higher surface with irregularities associated with excess energy.

Once the melting parameters are optimized, it is quite common to achieve densities above 99.9%. The residual porosities are then often of spherical morphology and result from the gas trapped in certain powder particles during atomization (Figures 1.39(a) and (b)). Porosity rates of around 0.05% have been reported in the literature for Ti-6Al-4V (Figure 1.39(c)) (Suard et al. 2015; Persenet et al. 2018). Densities close to 100% can thus be achieved by taking care to select powders devoid of porosities.

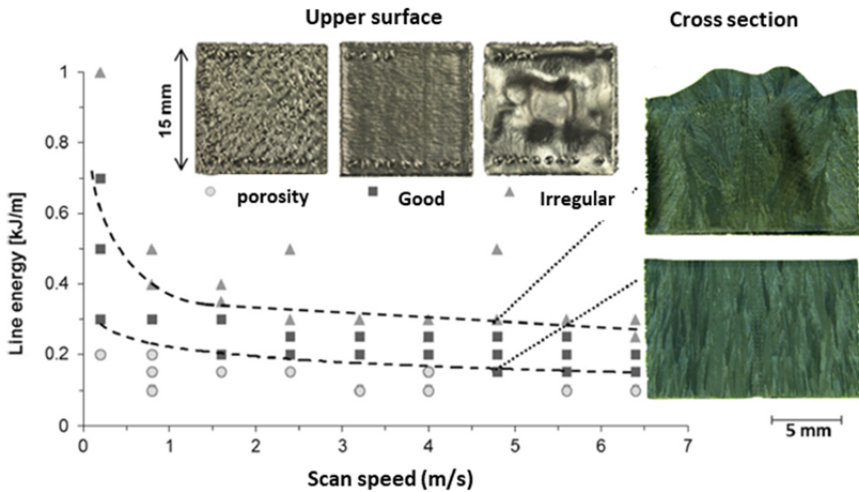


Figure 1.37. Mapping (linear energy ($E_l = P_0/V_0$), V_0) according to Juechter et al. (2014); view of the upper surface of Ti-6Al-4V cubes manufactured by E-PBF: parameters giving rise to porous samples (lack of energy) and optimized (>99.5% density) and “degraded” surfaces (excess energy)

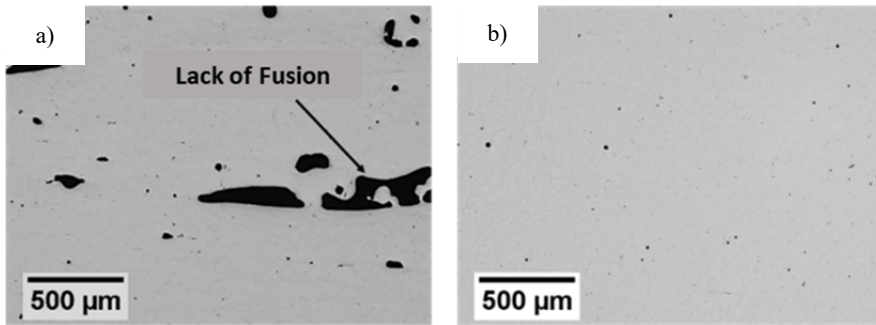


Figure 1.38. Optical micrographs of a) an E-PBF sample showing “lack of fusion” type defects and b) a dense sample with the presence of a few residual micro-pores (trapped gas)

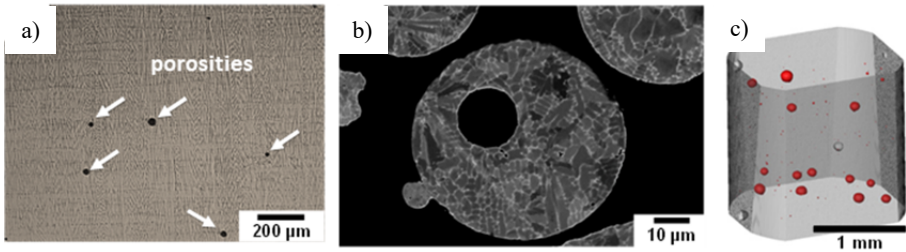


Figure 1.39. (a) Optical micrograph of an Inconel 625 E-PBF with spherical porosities. (b) Gas trapped in atomized powders of Inconel 625. (c) Three-dimensional reconstruction of a sample of Ti-6Al-4V manufactured by E-PBF, according to Tamas-Williams et al. (2015)

1.3.3.2. Evaporation of alloying elements

Extremely high temperatures can be reached in the molten pool and since the process takes place under vacuum, evaporations of alloying elements can occur. These evaporation phenomena can have two consequences on the quality of the material produced.

First, the selective evaporation of elements such as aluminum, zinc or magnesium with very high vaporization pressures can lead to heterogeneities in chemical composition. Thus, a study showed that for Ti-6Al-4V (Juechter et al. 2014), the aluminum content decreases after E-PBF manufacture and may vary from

layer to layer. This aluminum evaporation remains reasonable in the case of Ti-6Al-4V, because the standard is still met after manufacture by E-PBF. For other alloys such as TiAl (Juechter et al. 2018), preferential evaporation may be more significant and lead to changes in microstructures due to changes in processing temperatures. One solution then is to introduce an excess of volatile elements into the powders in order to obtain the desired composition once the manufacture is complete.

Second, evaporation phenomena are also responsible for the irregularities of the molten surface observed on the samples produced, especially at high energies. The vaporization of certain alloying elements will disrupt the flow of molten metal (recoil pressure) and sometimes lead to local accumulations of material giving rise to surface irregularities (Figure 1.40). If this local accumulation of material is greater than a layer thickness, the new powder layer will not make it possible to fill the lack of material. Ultimately, reproduced on several layers, this type of problem can interrupt the process. This phenomenon, which is detrimental to additive manufacturing, can be exploited for surface texturing with reflections using an electron beam (Dance 2005).

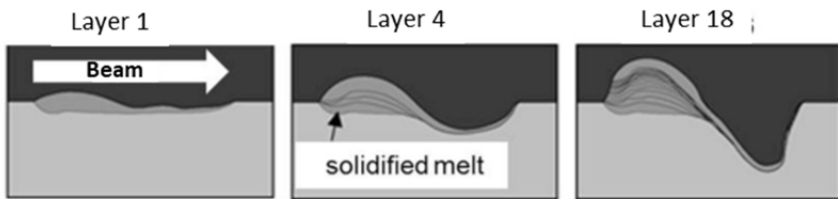


Figure 1.40. Simulation of the effect of evaporation of alloying elements on the flow of molten metal during the fusion of successive layers according to Körner (2016)

1.3.3.3. Defects from solidification

Other types of internal defects can be observed in parts produced by L-PBF (valid for L-PBF and E-BPF). This is the case with hot cracking phenomena (Figure 1.41(a)), including nickel-based superalloys with a large fraction of precipitates γ' . The 6000 and 7000 series aluminum alloys are the best-known cases. Micro-shrinkage can also form when pockets of fluid are trapped during the growth of secondary dendrite arms (Figure 1.41(b)). However, these problems are less severe than in L-PBF due to lower heating and cooling rates.

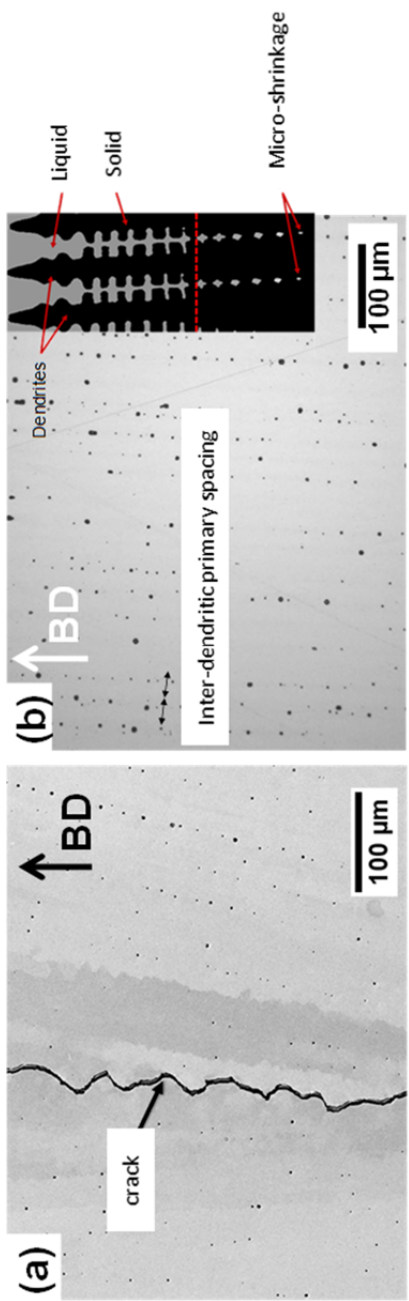


Figure 1.41. a) SEM micrograph of a crack in a non-weldable nickel-based superalloy developed by E-PBF. b) Micro-shrinkage on the same alloy. From Chauvet et al. (2018a)

1.3.3.4. Geometric defects

Dimensional consistency is always a challenge in additive manufacturing. Several factors are involved: the size of the powders, the size of the energy source, the change in volume associated with solidification and temperature variations and the presence of residual stresses. Figure 1.42(a) shows the delamination phenomenon that can impact the dimensions of E-PBF and L-PBF parts. Figure 1.42(b) illustrates a supported part exhibiting geometric variations linked to the change in dimensions induced by temperature variations. This has a direct consequence on the final dimensions of the manufactured parts.

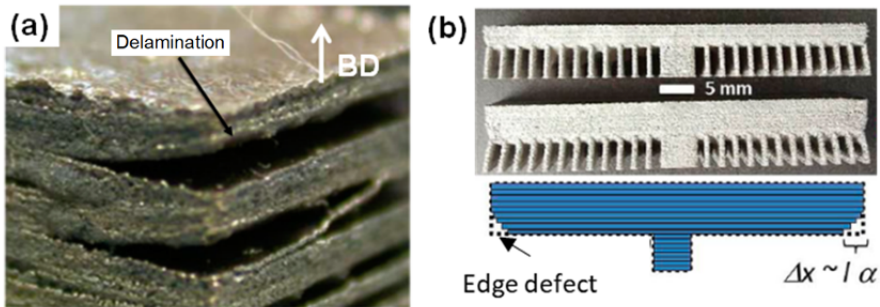


Figure 1.42. Geometric defects observed on elaborate parts in E-PBF:
 a) delamination between layers, extracted from Sames et al. (2016);
 b) distortion of edges, extracted from Körner (2016)

1.3.4. Other characteristics of the E-PBF process

1.3.4.1. Supports

The consolidation of the powder bed during preheating already described above also has the advantage of limiting the use of E-PBF supports, as the consolidated powder bed promotes the stability of the liquid metal and allows, to a certain extent, the dissipation of heat.

1.3.4.2. Surface state

The size distribution of powders, typically 45–100 μm , and the thickness of the layers, typically 50–150 μm , have a significant effect on the surface finish of E-PBF parts. The roughness R_a measured on raw parts is generally between 20 and 40 μm . As in L-PBF, this roughness deteriorates with the manufacturing angle relative to the vertical. Another consequence of surface roughness on raw parts is the poor prediction of actual manufactured sections, especially for thin structures (typically

beams constituting an architectural structure), leading to an inaccurate estimate of mechanical stresses $\sigma = F/S_0$ experienced by the part during external stress. Thus, raw parts, especially those of small dimensions, have surface roughness that does not allow the accurate estimate of mechanical stresses (Figure 1.43).

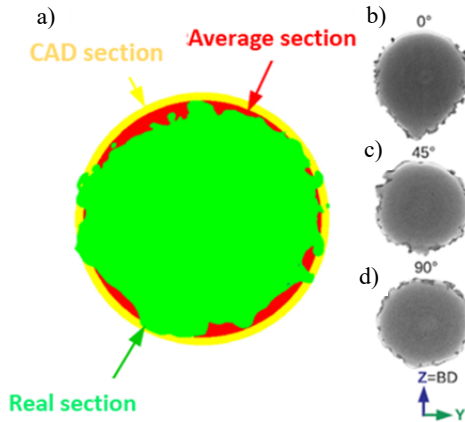


Figure 1.43. a) Illustration of the concept of the actual section. Samples of Ti-6Al-4V manufactured via E-PBF b) horizontally (0°), c) vertically (90°) and (d) at 45° (Persenot et al. 2018). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

Knowing the variability of the sections produced then becomes a major issue and approaches taking this into account have been proposed in order to anticipate these problems (Suard et al. 2015; Persenot et al. 2018).

1.3.4.3. Deformations and residual stresses

Unlike the L-PBF process, the E-PBF process, because of its preheating step followed by slow cooling after manufacture is complete (several hours), generally leads to lower residual stresses. This is confirmed in Figure 1.44, which presents a comparison of residual stress maps obtained on the same Inconel 718 part produced by L-PBF and E-PBF (Sochalski-Kolbus et al. 2015).

This absence of distortion is an advantage compared to the L-PBF process (with generally low temperature preheating), the parts of which have high residual stresses that usually require a thermal relaxation treatment before separating the parts from the build plate. Furthermore, preheating to a relatively high temperature can be beneficial for alloys sensitive to hot cracking phenomena. A high temperature thus makes it possible to limit the development of stresses of thermal origin.

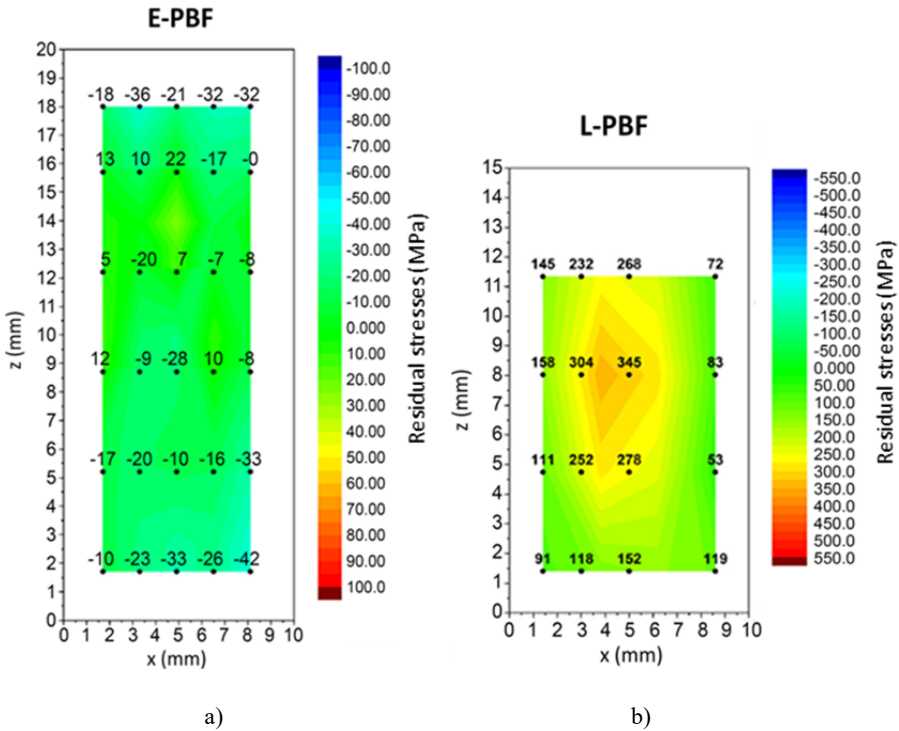


Figure 1.44. Residual stress maps in the direction perpendicular to the manufacture direction for a) E-PBF and b) L-PBF, taken from Sochalski-Kolbus et al. (2015). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

1.3.4.4. Thermal history: solidification and heat treatment in situ

The microstructures generated via E-PBF do not depend only on the solidification and cooling conditions, but can also be strongly impacted by the preheating acting as a *in situ* heat treatment. The temperature of this heat treatment varies with the material produced, typically 500°C for pure copper (Lodes et al. 2015), 700°C for Ti-6Al-4V (Juechter et al. 2014; Murr 2015) or up to 1,000–1,100°C for some nickel-based superalloys (Helmer et al. 2016; Chauvet et al. 2018a) or TiAl intermetallics (Juechter et al. 2018). This thermal history also explains why the melt-pool contours are much less visible in E-PBF microstructures than in L-PBF microstructures. A typical example of a point thermal profile of a part produced by E-PBF is given in Figure 1.45(a). After a first solidification, the material can be remelted several times, then subjected to thermal oscillations due to the fusion of the successive layers, which attenuate with time and with the build

height. A constant temperature is only achieved after manufacturing one to several dozen layers. Figure 1.45(b) shows thermal profiles measured at different depths during the melting of the top layer.

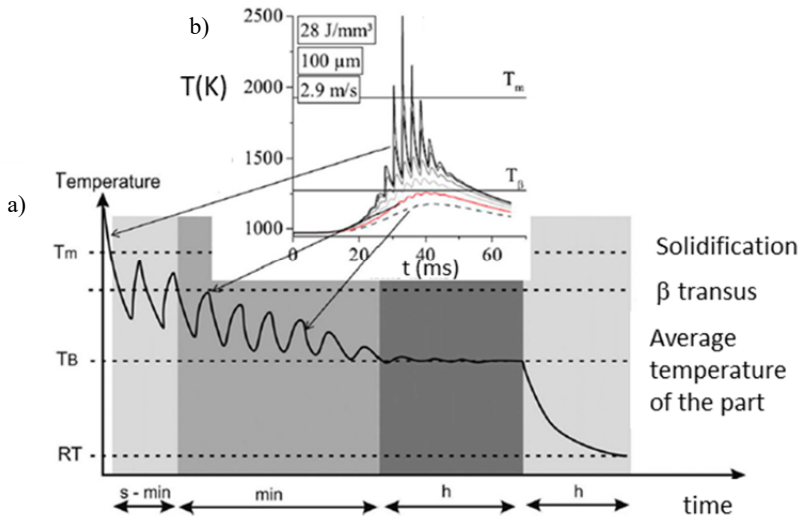


Figure 1.45. a) Thermal profile recorded at a fixed point on a part produced by E-PBF. Each oscillation is the addition of a layer. b) Thermal profile at different depths (100 μm increment) for Ti-6Al-4V. Demonstration of the differences in thermal cycles before and after fusion, according to Körner (2016). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

1.3.4.5. E-PBF microstructures

The microstructures produced by E-PBF are slightly coarser than those obtained by L-PBF, due to slightly slower cooling rates (10^4 – 10^5 $\text{K}\cdot\text{s}^{-1}$ vs. 10^5 – 10^6 $\text{K}\cdot\text{s}^{-1}$). The inter-dendritic spacing is then of the order of a few microns, while it is close to a micron in L-PBF. In all cases, this results in chemical segregation at very fine scales compared to what is observed in a conventional foundry. This aspect is doubly important: on the one hand, homogenization can occur, at least partially *in situ*; on the other hand, heat homogenization treatments make it possible to achieve the same metallurgical state much quicker.

As the chemical diffusion mechanisms become significant at the temperatures concerned, it is possible to observe microstructural changes such as homogenization or phase transformations in the solid state. Alloys exhibiting phase transformations

in the solid state are interesting because they offer the possibility of changing the texture induced by solidification. In the case of Ti-6Al-4V, the solidification gives rise to highly textured β phase columnar grains (orientation $\langle 001 \rangle$), but during cooling the β phase is transformed into α phase to give rise to an $\alpha + \beta$ microstructure known as the Widmanstätten structure (de Formanoir et al. 2016) (Figure 1.46). Even if the β columnar grains are still visible, the texture has been strongly modified by the $\beta \rightarrow \alpha$ transformation. The *in situ* heat treatment associated with preheating can also induce solid state precipitation, as in some nickel-based superalloys. Monitoring the size of the precipitates in the part then traduces the local thermal history experienced by the material (Chauvet et al. 2018a).

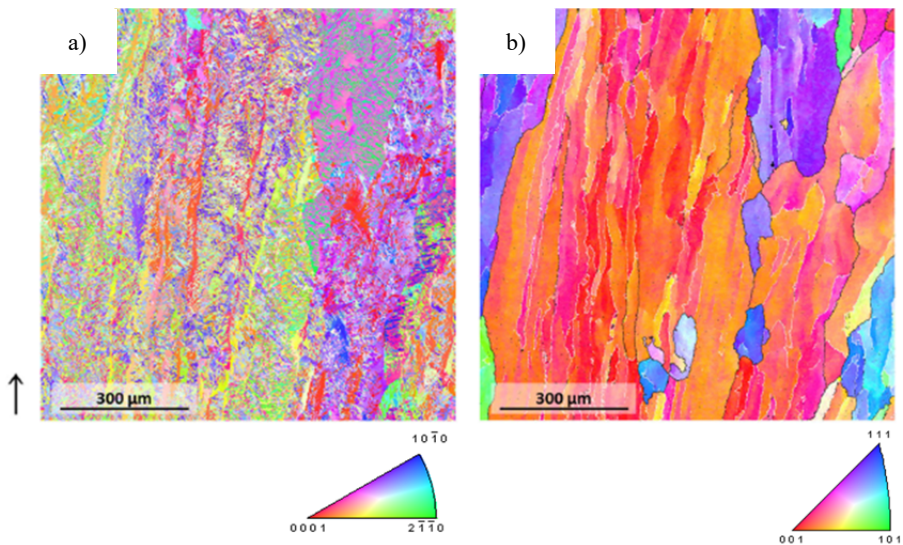


Figure 1.46. Electron back scattering diffraction (EBSD) crystal orientation maps on a sample of Ti-6Al-4V manufactured via E-PBF in the XZ plane (Z = direction of manufacture). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

COMMENT ON FIGURE 1.46.– a) Low-texture α slat microstructure with columnar structure in the background. b) Reconstruction of the highly textured microstructure of primary β grain with columnar grain: the $\langle 001 \rangle$ direction is aligned in the build direction Z. From de Formanoir et al. (2016).

The preheating step, specific to the E-PBF process, allows for an additional degree of freedom in order to control the amplitude of the thermal gradient as well as, most importantly, its orientation. Thus, by taking advantage of the different melting

methods, it is possible to modify the rate of solidification and the thermal gradient. This has been demonstrated in particular in nickel base superalloys (Helmer et al. 2016; Raghavan et al. 2017). Figure 1.47 (Dehoff et al. 2015) shows that it is possible, within the part, to locally control the texture and morphology of the grains. Monocrystalline structures can also be produced (Chauvet et al. 2018b) by thermal gradient control.

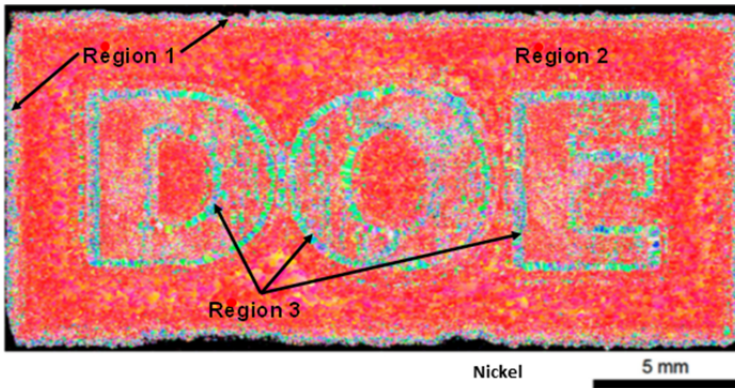


Figure 1.47. Crystalline orientation EBSD mapping with a Ni-based superalloy in the plane perpendicular to the build direction. Region 1 and region 3 consist of low-textured, equiaxed grains (no preferential orientation). Region 2 consists of textured columnar grains, oriented along $\langle 001 \rangle$. From Dehoff et al. (2015). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

1.3.5. Partial conclusion regarding E-PBF

The E-PBF process closely resembles its L-PBF cousin, although it is less popular among powder-bed melting processes and there are important differences between the two. First, only conductive materials (metals or alloys) can be produced by E-PBF. This is the case for the three most widely used alloys: TA6V for titanium alloys, Inconel 718 for nickel-based superalloys and ASTM F75 Cr-Co for Cr-Co alloys (Zhao et al. 2019). E-PBF can also be used successfully with pure copper, difficult to produce by laser because of its high reflectivity, certain non-weldable alloys, certain Mo or Nb refractory metals (Martinez et al. 2013) and even TiAl intermetallics (Juechter et al. 2018), industrialized for the production of turbine blades by the company Avio.

1.4. The deposition of matter via WAAM

In this section, we focus on detailing different additive manufacturing techniques based on arc welding processes with a supply of material in the form of wire. The objective is to give a broad overview of this technology both from the point of view of the process and the types of parts obtained and their properties.

1.4.1. Arc/wire additive manufacturing technologies

1.4.1.1. History

With regard to additive manufacturing technologies, the wire and arc additive manufacturing technique, based on a supply of material in the form of molten wire by an electric arc, has largely developed in parallel with powder techniques. The first remarkable applications of arc welding applied to additive manufacturing appeared in the years 1990–2000 with the establishment of a robust digital chain from design to manufacturing. During this period, a European project (Rapolac) proposed a technique based on a tungsten inert gas (TIG) welding process, a supply of wire, a 6-axis anthropomorphic robot and a rotator to position the torch in relation to the part being manufactured. It was about reducing the “buy to fly” for titanium parts. Subsequently, many developments were carried out at Cranfield University in England and at Wollongong University (Siminski 2003) in Australia on the use of an additive manufacturing welding process to produce large parts. Cranfield University, in particular, has extensively studied the problems associated with the supply of material in the form of a wire with a significant energy deposit: distortions, residual stresses and bead shape (Almeida 2012). In France, several university centers have been working on this technology since the 2010s (EC Nantes, ESTIA, University of Montpellier, Grenoble Alpes University, University of Toulon). Various scientific and technological advances have thus made it possible to market machines incorporating this type of process (Prodways, Gefertec). Despite relative technological maturity, problems remain:

- the positioning of the torch and the wire in relation to the part;
- the control of heat input and material deposition according to process parameters;
- the prediction and control of microstructures, distortions and residual stresses ensuring the mechanical integrity of parts;
- the ability of the digital chain to adapt to complex geometries.

In the following sections, different technological solutions used for wire/arc additive manufacturing are detailed, from the welding station to the machines.

1.4.1.2. *Different types of processes*

Different arc welding processes can be used depending on the material and the application for deposition:

- TIG welding (or micro-TIG);
- plasma welding (or micro-plasma);
- metal inert gas (MIG)/metal active gas (MAG) welding (with or without controlled short-circuit).

For the first two, the material must be supplied through an exterior device (push, push-pull) as close as possible to the weld pool. For a MIG/MAG, the fusible feeder wire is used as an electrode.

During TIG and plasma welding, an arc is created between a non-fusible electrode and the part using a generator. The flow of electrons and the plasma bring energy to the part and induce an FZ. The heat output is a function of the current I and the voltage U , which depend on the shielding gas. The current I is regulated by the welding machine and the voltage U is a function of the distance between the electrode and the part. Some stations allow the intensity to be controlled up to a few tens of amperes. The difference between TIG and plasma welding arises from the geometry of the nozzle, with the geometry of a plasma nozzle causing arc constriction, which involves more localized energy deposition. The distinction between MIG and MAG depends on the shielding gas: inert (MIG) or active with the presence of oxygen or CO_2 (MAG). Despite technological developments, it is more difficult to control the energy deposition in MIG/MAG welding than in TIG welding. An input configuration is shown in Figure 1.48. The shaded part corresponds to the weld pool created by the arc and the metal input.

As in TIG or plasma welding, the wire can be brought to the front but also to the side, the positioning of the wire in relation to the MP being an important parameter to control. In TIG, the supply of material therefore places a constraint on the orientation of the torch vis-à-vis the deposition. In MIG/MAG welding, the wire exits along the axis of the torch and the electrical parameters must be adjusted to ensure the fusion of the wire. Depending on the material, wire and gas diameter, a synergy (U , I) must be selected on the generator. Different transfer modes can then be encountered: spray, pulsed, globular or short-circuit (Air Liquide 2004). In additive manufacturing, the modes that generate the least energy are favored: pulsed and controlled short-circuit modes (mechanically, cold metal transfer (CMT) from

Fronius or, electrically, surface tension transfer (STT) from Lincoln). The combination of the relative scan speed of the torch V_0 and of the wire feed velocity V_f (Figure 1.48) and its diameter makes it possible to control the amount of material deposited. As an indication, the scan speeds in TIG welding are less than $300 \text{ mm} \cdot \text{min}^{-1}$ and in MIG/MAG welding less than $1 \text{ m} \cdot \text{min}^{-1}$. This results in much higher deposition rates ($\text{m}^3 \cdot \text{h}^{-1}$) in MIG/MAG welding.

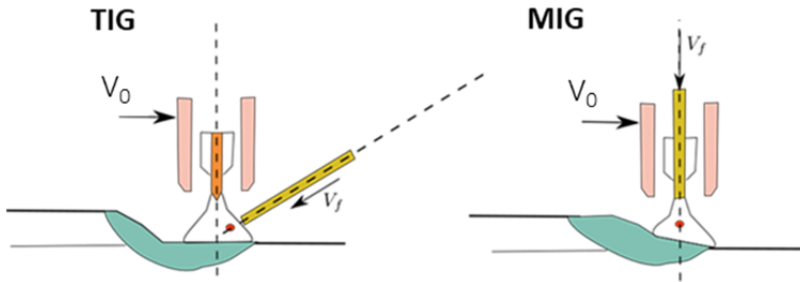


Figure 1.48. Additive deposit configuration with a TIG or MIG welding machine.
For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

1.4.1.3. Machines

Depending on the geometry of the parts to be manufactured, different technologies can be considered: (1) a 3-axis or 3+2-axis machining center or (2) anthropomorphic robots with turning and/or linear guide accessories. Examples of machines are shown in Figure 1.49.

The selection criteria for a machine are positioning accuracy, repeatability, speed and size of the working area. The welding process requires relatively slow feed rates and the precision of the deposition cannot be ensured below 0.5 mm . Machining centers have very good characteristics in terms of precision, repeatability and rigidity. They also allow a direct integration of the machining phase after the deposition phase. Robots are less precise in positioning, but allow for greater work volumes. The criteria for choosing a machine depend on the intended application. For a simple shape (extrusion in one direction), a 3-axis machining center is sufficient. For more complex geometries, the part will have to be reoriented and a 3+2-axis architecture will be necessary. Finally, for the production of complex surfaces for which the torch must be repositioned throughout the manufacturing process, 5-axis machines or robots will be preferred. Considering only the welding part, all the machines meet the criteria of rigidity, positioning and repeatability with regard to the precision induced by the process.

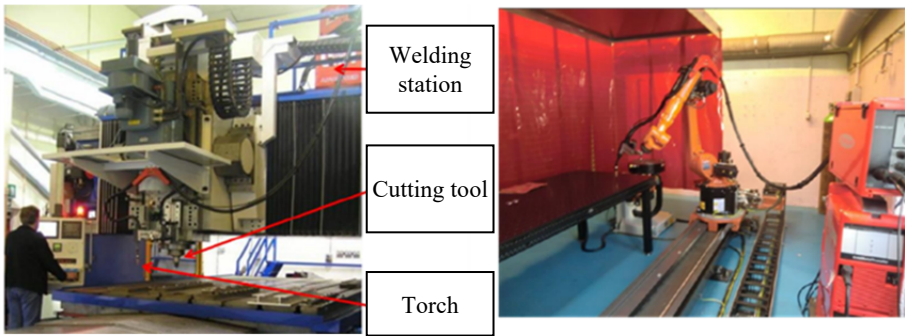


Figure 1.49. Positioning machines. Machining center (Cranfield University) and anthropomorphic robot with linear guidance (University of Montpellier)

1.4.1.4. The types of parts that can be produced using WAAM



Figure 1.50. Structures made with the WAAM welding process

Different parts produced with the WAAM process are shown in Figure 1.50. These are the following:

- 1) a light and complex wire structure made of 316L steel at the University of Delft (Figure 1.50(a)), which is difficult to machine;
- 2) an extruded part in titanium (Figure 1.50(b)) produced with a French Prodways machine in which we can clearly distinguish certain remanufactured functional surfaces;
- 3) a complex geometry, 250 mm high made of structural steel (Figure 1.50(c)), without arc interruption and with permanent reorientation of the torch throughout the manufacturing process to deposit the material tangent to the surface;
- 4) a remarkable example of an aeronautical structure (Figure 1.50(d)) produced through a collaboration between the company STELIA and the École Centrale de Nantes. This part combines several deposit difficulties at the intersections of geometries and reorientations.

1.4.2. WAAM process parameters

The most important research activities and industrial achievements in WAAM are carried out with arc welding processes of the MIG/MAG type. This is mainly due to the productivity of this process compared to the TIG process and the complexity of plasma-type processes. The parameters of welding processes are very well described in numerous works (Air Liquide 2004; Benoit 2015; Paillard 2017). What then differentiates WAAM from welding is the obligation of automation in WAAM (via robotization or a Cartesian bench).

1.4.2.1. Deposition parameters

As the WAAM process is derived from welding processes, mainly from the so-called “semi-automatic” MIG/MAG process, most of the parameters are therefore identical to those used in welding. These include the welding current I , the welding voltage U (these two parameters are controlled directly by the wire unwinding speed in WAAM MIG/MAG deposition using a synergic station), the diameter of the wire/electrode, the deposition rate, the nature of the shielding gas, etc. As in L-PBF and E-PBF, some of these parameters are grouped together as one of the most commonly used parameters: linear deposition energy E_l (equation [1.10]):

$$E_l \text{ (J/m)} = \frac{\text{Deposition current (I)} \times \text{Deposition voltage (U)}}{\text{Deposition rate (V}_0\text{)}} \quad [1.10]$$

In the case of WAAM MIG/MAG deposition, the deposition energy controls the mode of metal transfer in the arc:

- low deposition energy: transfer of metal in short circuit or short arc;
- medium deposition energy: transfer of the metal into globular or large drop;
- high deposition energy: transfer of metal by arc spray.

Low energy modes are favored for the realization of the contour walls of the piece while higher energy modes are used for the filling of internal zones. One of the essential parameters in WAAM (as in most metal additive manufacturing processes) is the deposition method. This, although little used in welding, has great influence on the microstructures obtained in the parts and therefore on the properties (generally mechanical) of use. This point will be discussed in section 1.4.3.

1.4.2.2. Optimization of the WAAM process

As mentioned earlier, deposition energy is an important parameter. Figure 1.51 shows this influence very well in the case of the deposition of a first layer of stainless steel 308Si (X2CrNiSi19-9) in protective argon gas.

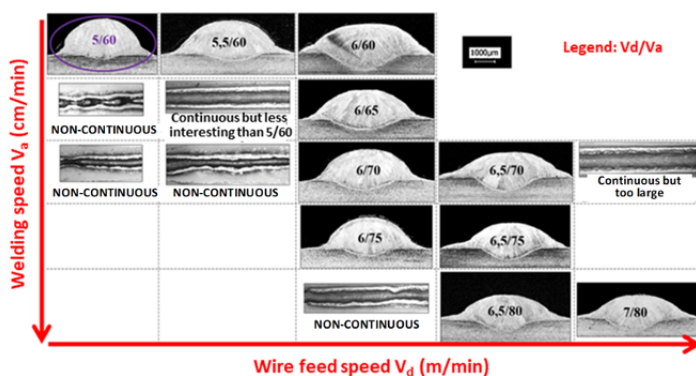


Figure 1.51. Influence of WAAM MIG deposition energy on the quality and the geometry of a deposited bead. In each macrography, the deposition rate-speed of advance coupling appears (Cuny et al. 2017)

An increase in the wire feed speed (V_f) and therefore E_l leads to wide and strongly penetrated beads (bottom right of Figure 1.51, for a wire speed of $7 \text{ m} \cdot \text{min}^{-1}$ and a deposition rate of $80 \text{ cm} \cdot \text{min}^{-1}$). On the other hand, for a lower energy (top

left, for a wire speed of 5 m/min and a deposition rate of $60 \text{ cm} \cdot \text{min}^{-1}$, the bead is little penetrated and is relatively large, leading to a high yield in the case of WAAM.

1.4.3. Use of materials

1.4.3.1. Overview

Most materials can be deposited by WAAM. The problems with the use of certain materials relate to the oxidation phenomena that can be generated during manufacture. Indeed, some materials are highly sensitive to oxygen contamination. This is the case with titanium and its alloys, stainless steels and zirconium and its alloys. For example, titanium alloys, which are the most sensitive, form a hard and fragile phase through oxygen diffusion, known as the alpha case. With this contamination taking place beyond 400°C , it is therefore necessary to increase the gas protection during deposition but also during cooling of the part. To do this, specific shielding devices can be used (in the case of stainless steels), but in the case of titanium and its alloys, the deposition should be carried out in a neutral atmosphere. WAAM equipment manufacturers therefore offer totally inert sealed robotic cells. This equipment increases production costs (several cubic meters of argon to be used each time the cell is opened). It is also possible to isolate the part being manufactured from the ambient atmosphere by waterproof bags (welding tents) or rigid boxes surrounding the part. These less expensive technologies are already used to weld titanium alloys that are very sensitive to O_2 contamination.

As in welding, the high cooling rates in WAAM can lead to the formation of martensite for low alloy steels. The presence of this hard and fragile phase can then lead to the formation of so-called “cold” or “hydrogen-induced” cracks. To avoid this, it is advisable to control the cooling rates of the parts through, as in the case of welding, preheating of the substrates and the parts during manufacture.

Finally, for materials sensitive to hot cracking, it is advisable to promote low energy deposits (CMT), to distribute the beads over the entire part (in the case of large parts) in order to reduce (below 150°C) the temperature of the deposits before building a new layer.

1.4.3.2. Input materials

Suppliers of filler metal for welding offer a large number of materials in the form of wires (see Chapter 2 of this volume). On the other hand, the grades offered are not always identical to the alloys used in subtractive manufacturing (machining,

etc.). Indeed, the composition of the supplied alloys is often more limited than that of the alloys that one wishes to assemble by welding:

- for low alloy steels, grades with identical mechanical properties are often low in carbon, enriched in volatile elements (manganese for example) and contain a higher rate of deoxidizing elements such as silicon;
- for stainless steels, the nature of the filler metal is often different from the base metal: austenitic stainless steels of type 304L are welded with filler metals of grade 308Si or 309;
- so-called precipitation-strengthened aluminum alloys (2000, 6000 and 7000 series) are assembled by welding with grades of 4000 and 5000 series to prevent hot cracking in the bead.

As a result, certain alloy grades are not in the catalogs of filler metal suppliers for welding: aluminum alloys of the 2000, 6000 or 7000 series and low and highly alloyed steels (stainless steel). Since it is relatively simple to obtain wires made from bars or billets, this problem can be solved quite quickly. Indeed, some companies (such as FSH Welding) can custom manufacture wires from metal blocks or bars (Doumenc 2020). In addition, some suppliers of filler metals for welding (such as voestalpine) supply varieties of filler metals in the form of wire specifically dedicated to metal additive manufacturing.

1.4.3.3. *Microstructure/process parameter relationships*

For WAAM, the beads are generally bulky (this is a high deposition rate process) and since it is liquid-phase metal deposition, the microstructure usually consists of large columnar dendritic grains. In addition, since the grains of the N pass germinate epitaxially on those of the N-1 pass, there may be granular continuity from one pass to another. For allotropically transformed alloys such as steels, it is possible to hope that the columnar coarse-grained microstructure can be regenerated by heat treating subsequent deposits. Figure 1.52 shows the influence of the manufacturing strategy in the case of TA6V walls made with 30 successive passes in WAAM TIG. The first wall (Figure 1.52(a)) is made by successive round trips (bidirectional strategy) without pausing between the layers. The build-up of heat in the wall due to the continuity of the deposit leads to an increasingly significant dilution (reflow of the previous layers) of the previous passes. The latter results in a lower wall which thickens during manufacture and has a large columnar prior- β grain (CBG) zone (50% of the height) compared to the equiaxed prior- β grain (EBG) zone.

The second wall (Figure 1.52(b)) was made with a unidirectional strategy and a 24 second pause between each layer. This method leads to less dilution and a higher and narrower wall in its upper part. In addition, the colonized grain zone only represents a third of the height of the wall. The last wall (Figure 1.52(c)) is made using a unidirectional method and a pause time of 120 s between each layer. It is the tallest and straightest, and the one with the best micro-structure (100% equiaxed grains) due to the longer cooling time between two layers.

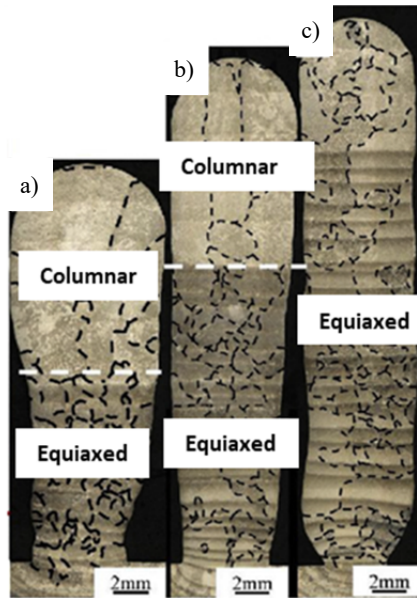


Figure 1.52. WAAM microstructures of a titanium alloy (Wang et al. 2019)

Another example of the influence of the deposition method on the microstructures obtained is given in Figure 1.53 for a 6061-aluminum alloy part made via WAAM MIG. The deposition is done unidirectionally (Figure 1.53(a)) or back and forth (Figure 1.53(b)), with identical deposition energies and pause times between layers. The microstructure of the unidirectional deposit has more columnar grains and a preferential orientation. This type of microstructure can be detrimental by promoting the propagation of hot cracks. In bidirectional deposition, an alternating, less textured microstructure is obtained, with more equiaxed grains and

alternating inclinations of columnar grains with respect to the direction of deposition. This microstructure is both more mechanically favorable and more resistant to hot cracking.

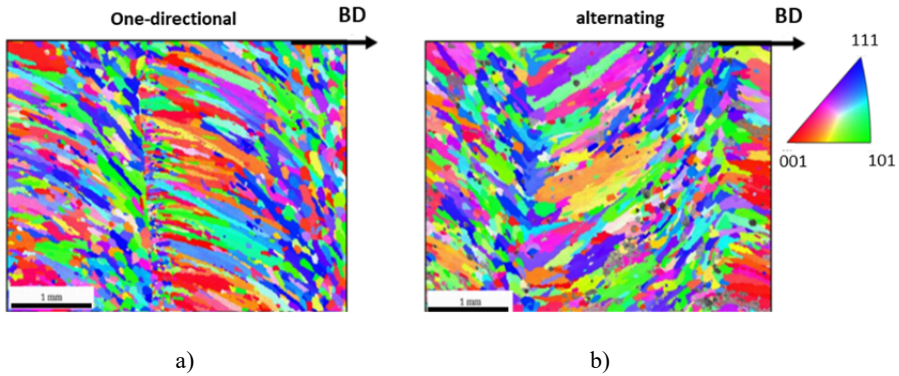


Figure 1.53. EBSD maps of WAAM microstructures on an aluminum alloy. Deposition strategy effect: a) same direction and b) alternating direction (Doumenc 2020). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

1.4.4. Residual stresses and distortions

Thermal depositions like WAAM lead to temperature and expansion gradients. As a result, parts manufactured by WAAM, as for welded assemblies, can deform. Figure 1.54 shows the effect of deposition energy on residual deformation. The first wall (figure 1.54(a)) was made in austenitic stainless steel (304L) in WAAM MAG (argon + 1.5% CO₂) with an average power of 3,150 W over the 10 layers deposited. An increase in power (4,860 W), the other conditions (gas, deposition rate, deposition strategy) remaining constant, leads to a greater deformation of the substrate (Figure 1.54(b)). In simple-geometry parts, it is easy to straighten them after manufacture. On the other hand, in the case of complex parts, it is recommended to distribute the metal deposits in such a way as to minimize the deformations.

If the substrate is solid (self-clamping) or strongly clamped, these temperature and expansion gradients can generate residual stresses in the part. Although these residual stresses are partly relaxed during manufacture by the heat treatments imposed during the subsequent deposits, they should be reduced or even eliminated because they can limit the fatigue life.

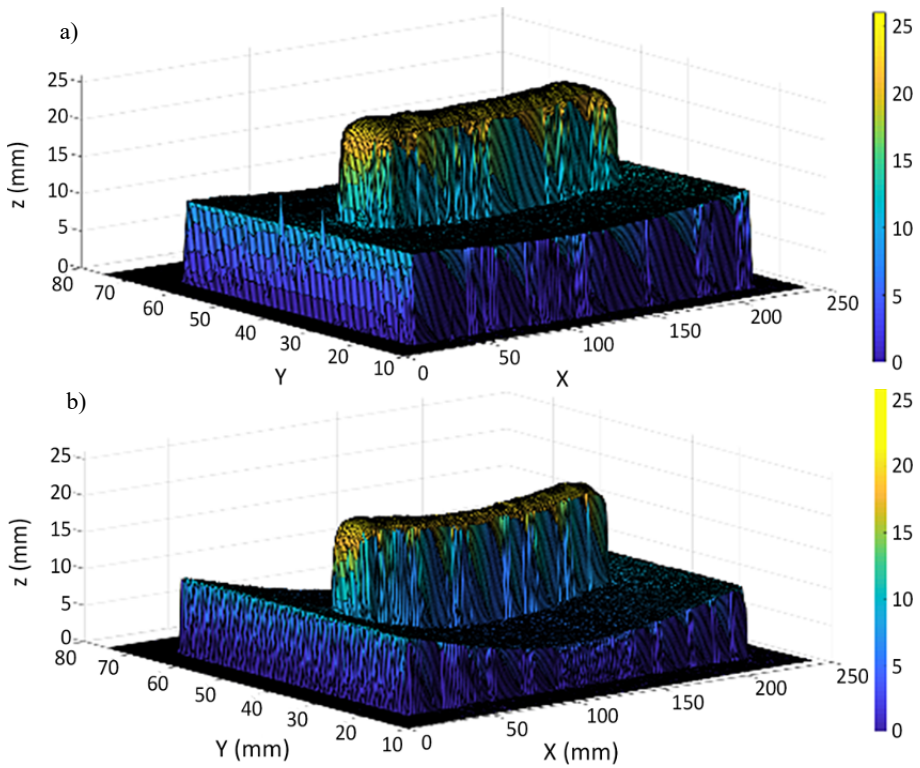


Figure 1.54. Measurement of deformations a) in the case of a low-energy deposition and b) in the case of a high-energy deposition (Guilmois 2020).
For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

1.4.5. Finishing process

Heat treatments (HTs) are often carried out after WAAM, with the aim of:

- relaxing the residual stresses (all materials);
- homogenization (austenitic stainless steels);
- solution heat treatment then tempering (precipitation hardening aluminum alloys).

In addition, the parts produced by WAAM generally have surfaces that are unsuitable for direct use. In most cases, the parts are therefore machined in order to obtain surface states identical to those obtained in subtractive manufacturing.

In order to be able to produce “near net shape” parts, studies are underway on the use of shielding gases that promote the wettability of the liquid metal via surfactant effects (Cuny et al. 2017). Initial results are presented in Figure 1.55.

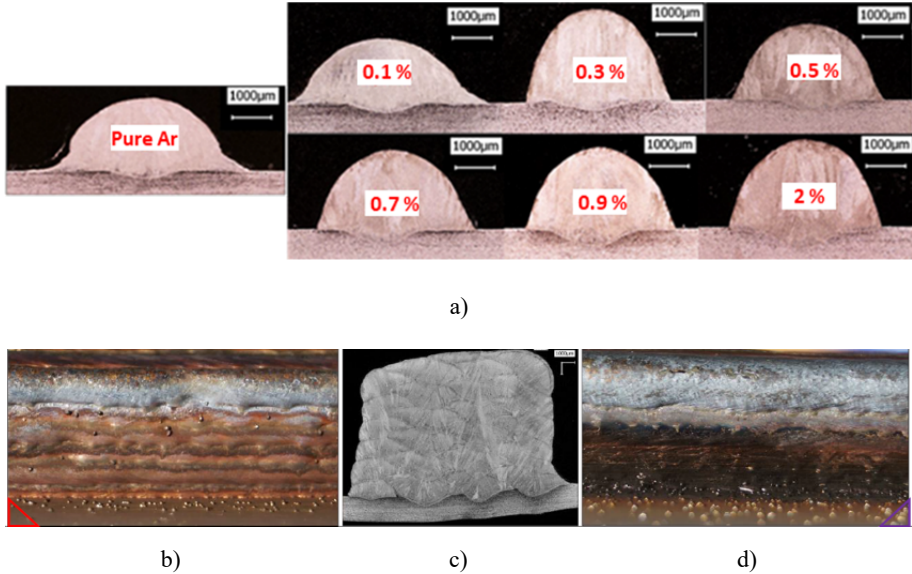


Figure 1.55. Influence of shielding gases in WAAM MAG. a) Effect of the O₂ content and c) macrographic section of a block made with the left side b) under pure argon and the right side d) under argon + 0.3% O₂ (Cuny et al. 2017)

Despite a supply metal rich in silicon (308Si), the influence of oxygen in the shielding gas on the shape of the deposited beads is relatively visible. It is then possible to obtain improved parts (Figures 4.8(c) and (d)) through the use of appropriate gas mixtures.

1.4.6. Digital chain: online control

1.4.6.1. General information on the digital chain in WAAM

The digital chain is the key to success for additive manufacturing in the context of the factory of the future. It must be able to adapt quickly in any shape. Several specific features of the process constrain the digital chain:

- the fact that wire additive manufacturing is done without support, which requires specific algorithms;

- the choice of the right process parameters to predict the shape of the deposit;
- the acquisition of a draft and the need for coupling with other finishing processes;
- the deposit rarely being in conformity with the initial CAD, in particular for complex shapes; it is then necessary to consider an online control.

In the following, an offline digital chain is detailed first, then the integration of control into the digital chain is discussed.

The digital chain of an arc-based AM process is presented in a simplified manner in Figure 1.56 (Hascoët et al. 2017). The solid lines are the interactions that exist during offline management, and the dotted lines are the feedbacks to be incorporated during online monitoring. The entry point is the geometry from a geometric modeler.

Geometry is normally imported in .stl format from CAD software. To generate the trajectories, different algorithms can be used depending on the filling type of the part. These algorithms resulting from material removal processes make it possible to adapt to complex shapes.

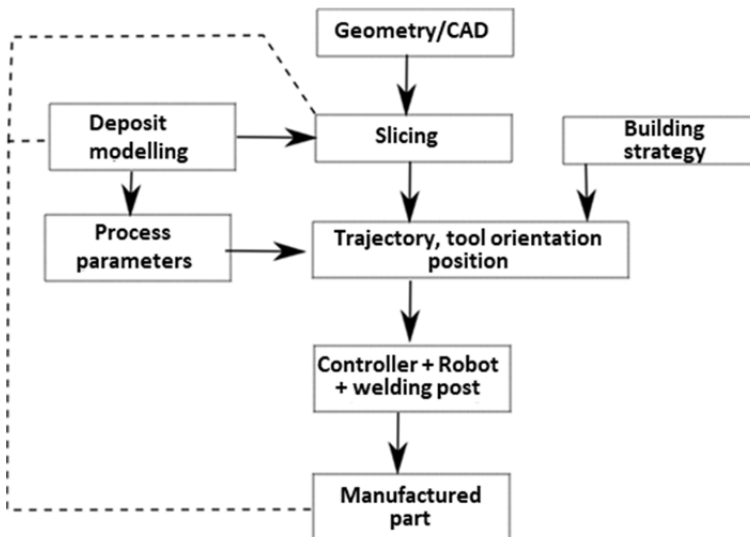


Figure 1.56. Digital chain for WAAM

1.4.6.2. *Manufacturing strategy and operating parameters*

As with other processes, the part is discretized into layers during slicing, then contours are created corresponding to the external geometry of the part. Filling then makes it possible to obtain a solid geometry (Ding 2015) with a large number of possible methods, such as in E-PBF and L-PBF.

Within the operating parameters, a distinction is made between parameters linked to the welding process (wire speed, synergy; see Chapter 3 of this volume) and those linked to torch movement (displacement speed, orientation, smoothing, etc.). Depending on these parameters and depending on the position of the bead, the width and height of the bead can be estimated. Following the slicing and the methods chosen, we then define waypoints and the orientation of the torch. For TIG and plasma welding, the position of the wire relative to the part places additional stress on the orientation. The waypoints are translated for the robot controller, which will in most cases control the welding generator.

When the geometry estimate of the deposit is good enough, the distance between the contact tip and the part remains constant during layering. For complex geometries, the energy transfer can be altered and the shape of the deposit may diverge, especially because deposit patterns are often identified for single beads. Under these conditions, one must be able to estimate and provide for corrections to the feed speed, the wire speed or the position of the torch in order to keep the part–contact tip distance constant. It is then necessary to integrate a feedback loop and an in-line control into the digital chain (Figure 1.56). Finally, various European projects (LASIMM² and KRAKEN³) have demonstrated the ability of the process to produce parts several meters in length with robot-based prototype solutions.

1.4.6.3. *In-line control*

In-line control is a topic under development, which should allow even greater integration of the arc-wire process in the context of the factory of the future. In order to ensure the good deposition of the material, the deposited geometry can be measured just after the deposition. This approach allows parametric corrections to stabilize the process (Heralić et al. 2012). On the other hand, energy transfers can be appreciated in real time during manufacture by measuring tension or observing the size and shape of liquid areas from rapid imaging.

2 Available at: <http://www.lasimm.eu/index.html>.

3 Available at: <https://www.krakenproject.eu/>.

1.4.7. Conclusion

In conclusion, the advantages and disadvantages of the WAAM process are listed in Table 1.3.

Advantages	Disadvantages
Low installation cost	Lower resolution
High deposition rate (10 kg/h)	Rippling in the manufactured surface
Excellent hygiene and safety	Significant residual stresses
Compatible with other processes	Complex implementation on bimetals
Can be used with any weldable material	Manufacturability of left shapes (double curvature) not yet demonstrated

Table 1.3. *Advantages and disadvantages of the WAAM process*

1.5. Emerging processes

1.5.1. Indirect fabrication by selective laser sintering and infiltration

1.5.1.1. General overview

Initially, SLS was an additive process, patented in 1986 (Deckard 1986), used primarily for printing objects made from polymer materials. Its principle is similar to that of the selective melting process on a bed of metal powders, but the laser used is different (CO₂ laser; $\lambda \sim 10.6 \mu\text{m}$). Like all additive technologies, the object to be produced is first modeled by CAD and sliced into two-dimensional sections with a thickness of around 100 μm . The layers are printed successively according to the diagram in Figure 1.57 after spreading using a roller or a scraper. An infrared laser then makes it possible to selectively melt the powder according to the 2D sections of the CAD model, which makes it possible to transform each section into a 3D solid. Usually, the powder bed is preheated and the laser locally supplies only the thermal energy necessary to bring the polymer to its melting temperature. The SLS process can also be used to build all-metal parts. In the latter case, we speak of indirect SLS since it takes several steps to obtain a fully metal part. The first step involves printing a porous preform through SLS sintering of a predominantly metallic powder mixed with polymer particles, which act as a binder (~10% by volume of the mixture).

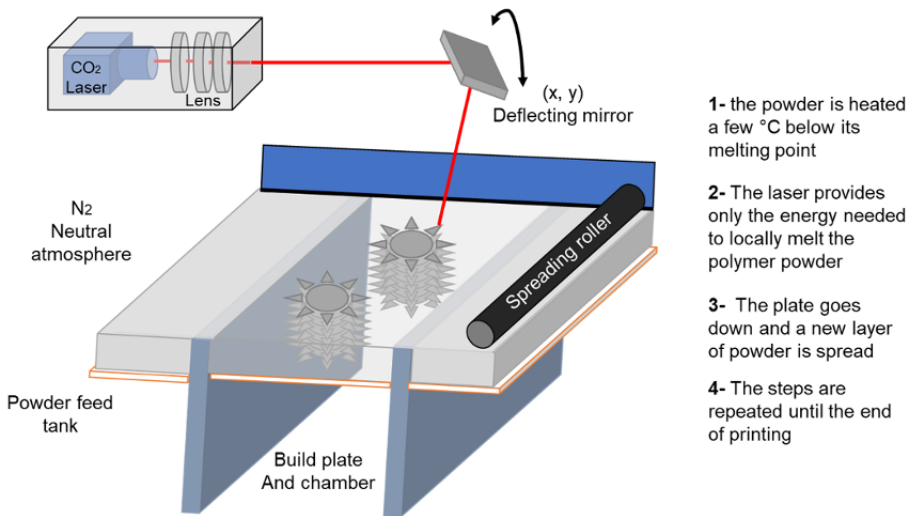


Figure 1.57. SLS: powder (polymer base) unmelted by the laser naturally serves as a support for the following printed layers

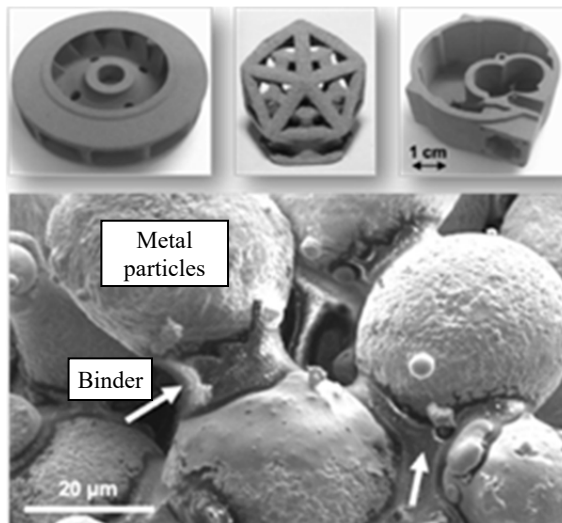


Figure 1.58. Examples of porous preforms (40–50% volume) obtained after SLS and SEM image (secondary electrons) showing the atomized metal particles bound by the polymer to the interfaces symbolized by the arrows

A metal/polymer composite preform is then obtained, which contains approximately 50% by volume of porosities (Figure 1.58). It is stiff enough to be handled, but too fragile to be used directly as is. The preform must therefore undergo heat treatment during which the polymer binder is first displaced in the form of gas (debinding), then the preform is infiltrated by a second alloy whose melting point is necessarily lower than that of the preform. In this case, the metallic infiltration of the porous preform takes place by capillary action as soon as the infiltration conditions are satisfied. At the end of this infiltration treatment, we then obtain an all-metal composite part consisting of the metal base and the infiltrated alloy.

1.5.1.2. *Theoretical ideas: capillary infiltration*

Capillarity is a phenomenon consisting of the ability of a liquid to rise along a surface by freeing itself from the force of gravity. From a physical point of view, it can be described by the interactions occurring at the interfaces between two immiscible liquids, between a liquid and an atmosphere (vapor) or between a liquid and a solid.

In the example in Figure 1.58, a porous metal alloy preform should be infiltrated by a second molten alloy. This generates a triple interface between a liquid (here a metal alloy), an atmosphere and a solid (network of spherical particles). During infiltration, this interface is mobile in the porous network up to an equilibrium level. The best-known example of capillarity is the rising movement of coffee on a lump of sugar. We can then make the analogy with a liquid metal with a low melting point, which infiltrates a solid metal skeleton with a higher melting point. The capillarity is dependent on the surface tensions which set the wettability of a solid surface by a liquid, expressed by the wetting angle (θ) at the solid/liquid interface. This angle is linked to the different solid/liquid ($\gamma_{s/l}$), liquid/gas ($\gamma_{l/g}$) and solid/atmosphere ($\gamma_{s/g}$) interfacial tension couplings via the Young–Dupré equation (Figure 1.59). Angle θ is characteristic of the wettability of a fluid, hence its ability to spread over a solid surface. The lower this angle ($<90^\circ$), the more the liquid will wet the surface.

The spread of a metallic liquid in a preform is mainly governed by the antagonistic effects of capillarity and gravity. The Lucas–Washburn model gives an analytical solution (Washburn 1921). The contribution of capillarity is governed by the surface tension between the infiltrant, the infiltrate and the contact angle between the two materials.

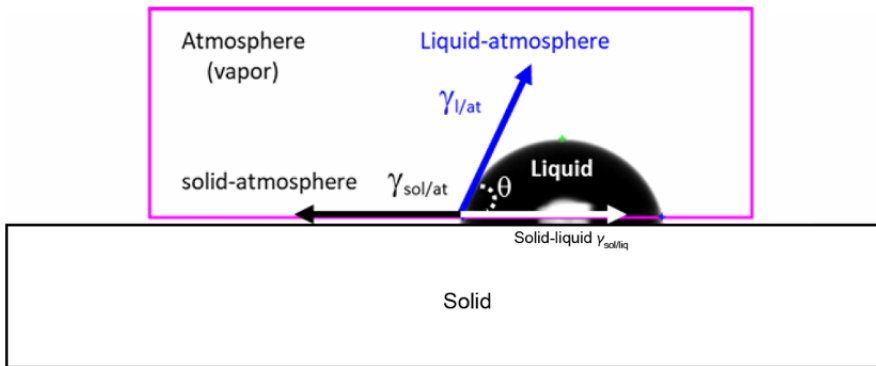


Figure 1.59. Descriptive image of a drop of liquid forming a wettability angle θ on a horizontal solid surface. Young–Dupré equation: $(\gamma_{s/g}) = (\gamma_{s/l}) + (\gamma_{l/g}) \cdot \cos \theta$. For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

1.5.1.3. Examples of bimaterial SLS parts produced by infiltration

At present, indirect SLS-type processes do not yet compete with direct additive processes and are still in the laboratory stage. They do, however, make it possible to combine metallic materials, with many perspectives in the range of materials considered and in the evaluation of their functional properties. The indirect SLS process is fairly well mastered for the manufacture of steel/bronze composite parts (see Figure 1.60). The pieces obtained are dense ($\sim 8 \text{ g} \cdot \text{cm}^{-3}$) and their mechanical properties are acceptable (Kruth and Vandenbroucke 2005).

Some work has been done to lighten the parts produced and extend the range of materials to lighter alloys or composites such as aluminum alloys. The infiltration of an aluminum alloy into a preform itself based on aluminum is a real technological challenge, because aluminum alloys all have relatively close melting points, which quickly destabilizes the preform during high temperature infiltration by another aluminum alloy and degrades the dimensional stability of the infiltrated part.

Sercombe et al.'s (2003) work helped us to stabilize aluminum-based parts (Al 6000 and Al 2000) in obtaining preforms. These had been infiltrated by a whole range of aluminum alloys with a lower melting point ($T_f < 600^\circ\text{C}$). Many variants have been tested to improve the quality of the interfaces (addition of Sn or Pb to improve the wettability of the infiltrant). Despite all this, no alloy allows satisfactory mechanical resistance, but the stability of the parts is preserved (Figure 1.61).

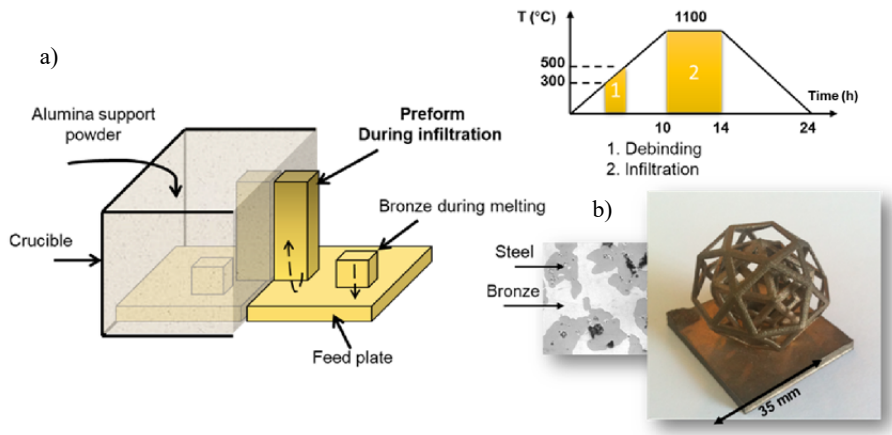


Figure 1.60. a) Bronze infiltration cycle in a steel/polymer preform. The bronze infiltrates through capillary action into the preform after debinding. b) Photograph of a steel/bronze architectural piece and associated micrograph

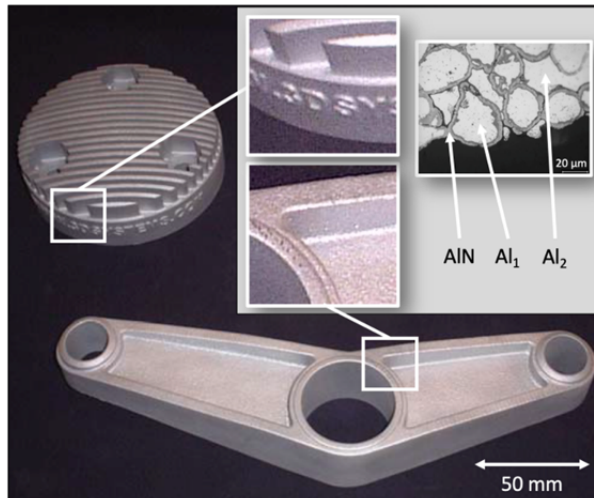


Figure 1.61. Examples of infiltrated aluminum parts (Sercombe 2003); in the insert, a typical microstructure showing the nitrided (AlN) interfaces between the aluminum base (Al₁) and the infiltrant (Al₂). Dimensional stability and level of detail are preserved

In these studies, the removal of the polymeric binder takes place in an oven under a nitrogen atmosphere (N_2). Nitrogen is important for the infiltration step because it allows nitriding of the interfaces (formation of aluminum nitride AlN) of the porous preform and creation of a rigid skeleton (ceramic type) whose thermal stability is superior to that of an aluminum alloy (Sercombe 2003). The nitrided free interfaces of the skeleton, of the AlN type, also promote infiltration by capillarity compared to the native oxide of the Al_2O_3 type (alumina) due to improved wetting: the wetting angle of aluminum in AlN is lower than that of aluminum in Al_2O_3 (Rosazza Prin et al. 2001). The infiltrated Al parts have mechanical properties, which depend on the nitriding time of the free interfaces before infiltration and on the nature of the infiltrant. Among all the infiltrants used (Al, AlSiMg, AlMg, AlCuMg), the rate of porosity and the weakness of nitrided interfaces generate at best mechanical strengths similar to those of aluminum, but associated with very low ductility (<1 % elongation).

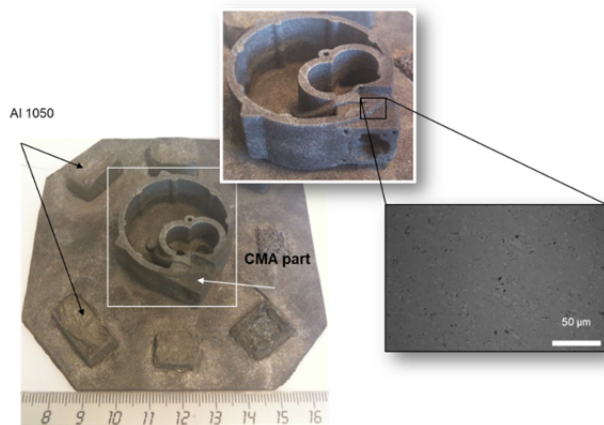


Figure 1.62. Complex metallic alloy AlCuFe-(B)/polymer composite preform debonded then infiltrated at 750°C with an aluminum 1050 alloy. After infiltration, the microstructure does not have the characteristics of a bimaterial (CMA/Al) composite and the part is made up of complex metal phases of the AlCuFe system

Recent work has explored other types of aluminum alloys to avoid the nitriding step, with the objective of improving the ductility of infiltrated parts. These are near-crystalline complex metallic alloys (CMAs), which have a significantly higher thermal stability than conventional crystalline aluminum alloys. These CMAs are perfectly ordered and their quasi-crystallinity comes from their quasi-periodic atomic arrangement (Shechtman et al. 1984). They exhibit singular properties, such as poor thermal and electrical conductions when they are made up of metallic

elements (e.g. Al, Cu, Fe) or good tribological properties, with low friction coefficient values (Dubois et al. 1991) associated with high resistance to wear. They also have good corrosion resistance and low wettability (low adhesion). On the other hand, they are too fragile to be used in the form of massive materials. Figure 1.62 shows a CMA AlCuFe-(B) part infiltrated under a primary vacuum with 1050 aluminum (99.8% Al). In conclusion, the indirect SLS process associated with metallic infiltration is still, although promising, a new field of exploration where technological obstacles remain. Applications on light alloys can only emerge when the process makes it possible to manufacture parts that are less brittle and more dimensionally stable, with a complementarity to direct processes.

1.5.2. Indirect manufacture via metal binder jetting

1.5.2.1. Principle of operation

As with SLS, metal binder jetting (MBJ) is an indirect additive manufacturing process that takes place in two main stages (Figure 1.63). First, printing gives the part its shape. This involves successively depositing layers of metallic powders onto which an ink (including a binder) is selectively projected to form a green part. Second, sintering consolidates the material and gives the part its final properties. The green part is brought to a temperature below the material melting point, but sufficient to activate various atomic diffusion mechanisms. For MBJ, this results in a densification of the part and, incidentally, an almost homogeneous linear dimensional shrinkage of approximately 14–16% (Figure 1.64).

Figure 1.65 shows the complete production cycle and the necessary intermediate steps between printing and sintering: “curing” (baking to harden the binder and make the parts easy to handle), “depowdering” (extraction of the green parts from the powder bed), “laying” (arrangement of the parts on support plates) and “debinding” (elimination of the binder).

This is comparable with the metal injection molding (MIM) process. The sintering step is similar (dimensional shrinkage, thermal cycle) as are the microstructures and mechanical properties obtained. However, differences exist with respect to the binder percentage (12% m in MIM, 2% m in MBJ) and the final roughness of the parts (Rt of 30–100 μm with “strates” for the MBJ, 20–35 μm for the MIM). As in MIM, depending on the size or shape of the part, a sintering support may be necessary to prevent sagging of the material or to guide shrinkage (Figure 1.66).

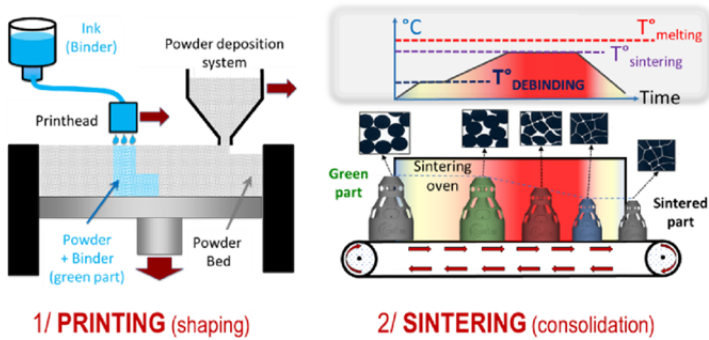


Figure 1.63. Principle of the MBJ process. For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip



Figure 1.64. Illustration of dimensional shrinkage on two MBJ parts. For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

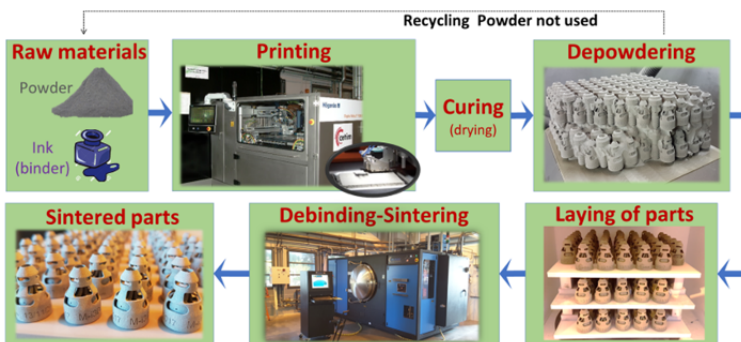


Figure 1.65. Complete MBJ cycle

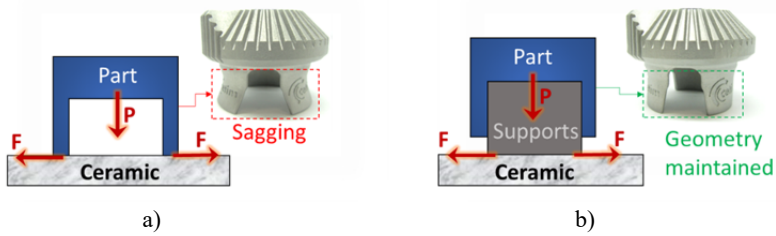


Figure 1.66. Configurations a) without and b) with sintering supports

1.5.2.2. Strengths and limitations

The absence of supports connected to the part reduces rework operations and opens up new design possibilities (e.g. the possibility of integrating mobile elements into the part). Compared to other additive manufacturing processes, the as-built roughness may also be better and more homogeneous (R_t of 30–100 μm in MBJ, 100–250 μm in L-PBF). Finishing operations are thus simplified.

The use of printheads (Miyanaji et al. 2018), the technologies for which are derived from printing, makes it possible to consider industrial systems comprising several tens of thousands of nozzles (compared to a few hundred today). The productivity potential is therefore very high, especially since the absence of melting in the process allows printing at room temperature (without inerting or vacuum constraint), at the same time facilitating the increase in production volume (Calves et al. 2018). However, MBJ is not suitable for all types and sizes of parts. Due to the high dimensional shrinkage and high sintering temperatures, the larger the part, the more difficult it is to maintain tight tolerances. Controlling the dimensional shrinkage is a major challenge; in addition, advanced knowledge in powder metallurgy is necessary to ensure real control of the process, in particular the sintering behavior (Bouvard et al. 2018; Mostafaei et al. 2019). Today, the mass of built parts rarely exceeds 400 g, with tolerances in the range of 0.1–0.2 mm at best. Representative examples of MBJ parts are shown in Figure 1.67.



Figure 1.67. Examples of parts manufactured using MBJ

1.5.2.3. *Materials*

To date, the number of grades available is still limited. Two stainless steels (316L, 17-4PH) are commonly used and some manufacturers offer titanium (TiAl6V4), Inconel 625 and 718 or MAR-M247. Copper, certain steels (X40CrMoV5, 42CrMo4) and cemented carbides (WC-Co) are being developed (Enneti et al. 2018; Mostafaei et al. 2019). Ultimately, the materials used in MIM as well as hard materials (carbides and ceramics) will be accessible. Materials with poor weldability can also be used. Compared with MIM, aluminum alloys are an exception and it is difficult to foresee frequent use in MBJ.

1.5.2.4. *Perspectives*

Some applications are currently in production for small, complex parts. The still significant productivity potential of MBJ, combined with the absence of part-related supports, will increase the number of economically viable applications in the future, and this with larger series, larger parts and new fields of activity. The process can be complementary to processes such as MIM and used with series where the cost of a mold or tooling is not profitable or parts where additive manufacturing can bring added value (customization, texturing 3D, shapes). Stakeholders, materials and machines, few in number today, are currently growing.

1.5.3. *Direct manufacture in the solid state without melting*

1.5.3.1. *Cold spray additive manufacturing*

The cold spray process is a method of depositing by projection of micrometric powders (10–100 μm). The material is supplied in an open environment by means of a De Laval nozzle device, which uses a hot inert gas under pressure (nitrogen, helium) to propel the powder at supersonic speed (Fauchais et al. 2013; Cavaliere 2018). Particle velocities can vary from 300 to 1000 $\text{m}\cdot\text{s}^{-1}$ depending on the nature, temperature and pressure of the gases; the expansion ratio; the nozzle shape; and the characteristics of the powders (size and density). The powder acceleration is produced in the diverging zone of the nozzle and the substrate is placed between 1 and 4 cm from the nozzle outlet. Two types of cold spray device can be used depending on the maximum pressure of the gas and the position of the powder injector upstream or downstream of the nozzle neck (Figure 1.68). Low-pressure devices (<1.0 MPa, 200–500°C) are less expensive and much more manageable, but are limited to very ductile materials. High-pressure devices (<6.0 MPa,

200–1,000°C) give higher impact velocities to propel less ductile materials, improving the adhesion, cohesion and compactness of the sprayed layers.

Moreover, this additive process is based on a supply of material in the solid state which is particularly suitable for ductile metals (Al, Cu, Zn, Ti, Ni, Ag, Ta) and a few alloys (Fe, Ni, Ti bases). Each powder particle is subjected to severe plastic deformation depending on its speed and temperature on impact, and also during the successive impacts of other particles during the increase in thickness.

For each material, it is necessary to reach a minimum critical speed to achieve attachment of the powder by activating an adiabatic shear phenomenon at the interface (Assadi et al. 2016). The movement of the nozzle allows the material to be removed in the form of beads with a width of around a centimeter. Thanks to a fairly homogeneous and directional material jet, material overflow at the periphery of the bead remains very limited. These characteristics make it possible to quite precisely control the areas to be covered, generally dispensing with a specific surface preparation and the masking of adjacent areas. The overlapping of these beads allows the production of thick layers ranging from a few tens of micrometers to several centimeters depending on the materials and the nature of the targeted applications: coating, repair, addition of function (bossage) or additive manufacturing.

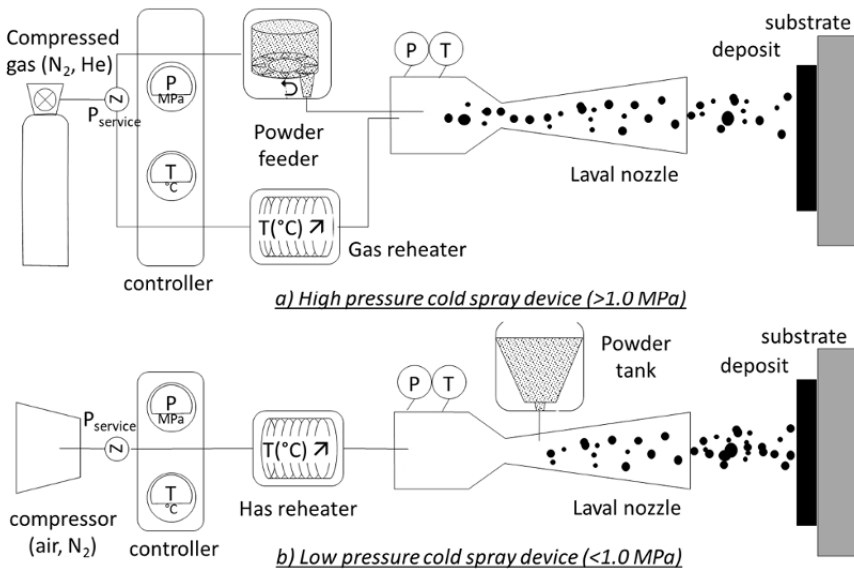


Figure 1.68. a) High pressure and b) low pressure cold spray devices

Compared to hot spraying and additive manufacturing by fusion, cold spraying preserves the powder in a solid state and limits overheating. The microstructures obtained are generally dense and free of oxides (high thermal and electrical conductivities) with strong adhesion without a dilution zone. The residual stresses are generally in compression. Powder flow rates of $5\text{--}10\text{ kg}\cdot\text{h}^{-1}$ with yields of around 80% result in high construction rates. The cold spray is therefore of interest for its productivity. On the other hand, low ductility and anisotropy of the properties may require a thermal after-treatment. The process also has limitations in the case of materials with low ductility and substrates sensitive to erosion.

Cold spray additive manufacturing allows the rapid production of parts or blanks without size limits and without pronounced thermal effects (Pattison et al. 2007). The possibilities are varied, but the details of the admissible morphologies are in the centimeter or millimeter range depending on the methods and the machining possibilities. The spatial distribution of the material jet and the overlap phenomena govern these shapes. The morphology of a single bead depends on the kinematic and mass parameters (powder flow rate, speed, nozzle orientation and notch). A thick bead has a triangular shape (see Figure 1.69(a)). This characteristic morphology results from the speed and the number of particles at the center of the jet. At the periphery, the particles adhere less because they are slower, less numerous and have deviated paths (Figures 1.69(b) and (c)). The edge effect then amplifies the incidence effect. Thus, the kinematic and mass parameters can be chosen to attenuate or promote these geometric effects which lead to beads with rather straight or rounded profiles.

Different shape concepts are currently achievable in cold spray additive manufacturing (CSAM) (Figure 1.70). There are many possibilities and they can be combined depending on whether it is a translational support plate or a rotating support mandrel. The support, sacrificial or not, can be straight or curvilinear or have a pre-machined profile. Likewise, multi-material components produced by CSAM are possible, including materials with gradients.

Different examples are given in Figure 1.71: (1) walls obtained on a fixed plate with tessellation or sweeping (Figure 1.71(a)), (2) fittings and stiffeners for aeronautics (Figure 1.71(b)), (3) self-supporting parts combining translational and rotational processes with machined or non-machined areas (Figure 1.71(c): on the left, an automotive part, on the right, cam coupling for naval equipment), (4) nozzles representative of axisymmetric mono- or bimaterial parts after machining of internal channels (Figure 1.71(d)) and (5) an axisymmetric component composed of two materials with different magnetic properties (Figure 1.71(e)).

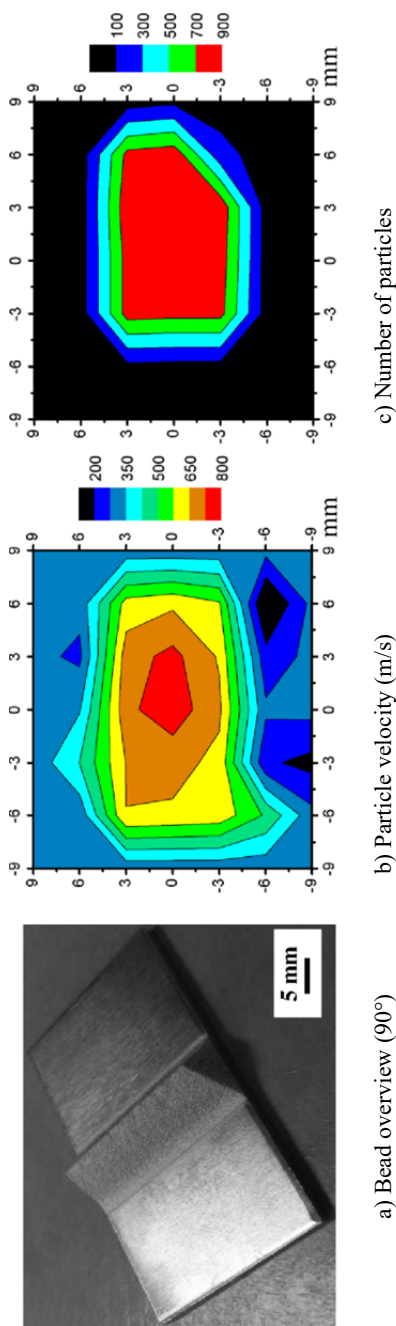


Figure 1.69. a) Unit bead (A1 050); b) and c) spatial maps of the jet. For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

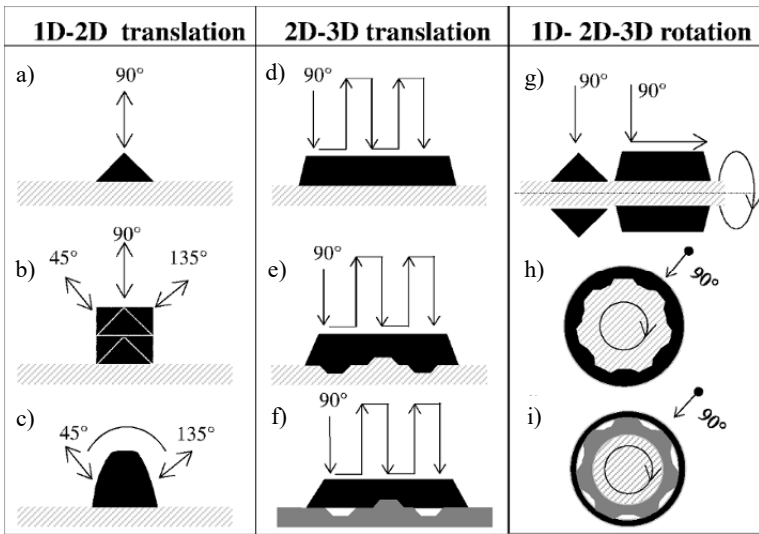


Figure 1.70. Concepts of shapes in CSAM: a) unit bead; b) wall by tessellation; c) wall by sweeping; d) single layer; e) single layer, machined plate; f) single layer, machined plate, partially sacrificial plate; (g–i) same as (a–f) for cylinders

1.5.3.2. MELD additive manufacturing

MELD is an additive process for metal deposits in the solid state developed over the past ten years by the company Aeroprobe (Aeroprobe Corporation 2016) and marketed by its subsidiary MELD Manufacturing since 2018. It has many similarities with the friction-based non-additive processes, such as friction stir welding (FSW) and additive manufacturing using friction stir additive manufacturing (FSAM). Both cases are solid-state thermomechanical processes allowing the assembly of metal parts without melting and without adding material (Griffiths et al. 2019). The materials used are in the form of powders or solid metal bars. Figure 1.72 shows the MELD principle in the construction of multiple metal layers on a substrate through a hollow rotary tool under uniaxial pressure. The metallic layers are formed successively as the hollow tool (or the substrate) moves. The temperature increases and the mechanical stresses imposed by the process ensure both a strong metallurgical bond and adhesion between the layers, while avoiding the formation of porosities (dense deposits).

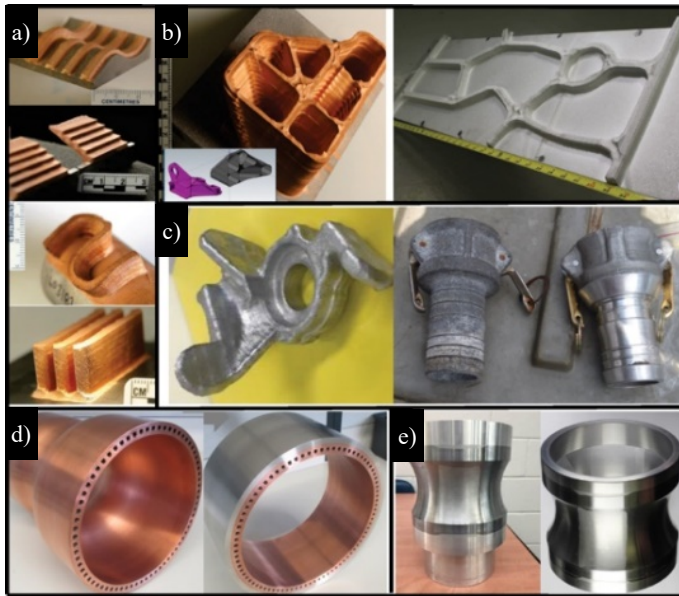


Figure 1.71. Examples of cold spray additive manufacturing parts produced by: (a, b, e) CNRC (Canada), (c) Spee3D (Australia) and (d) Impact Innovations (Germany)

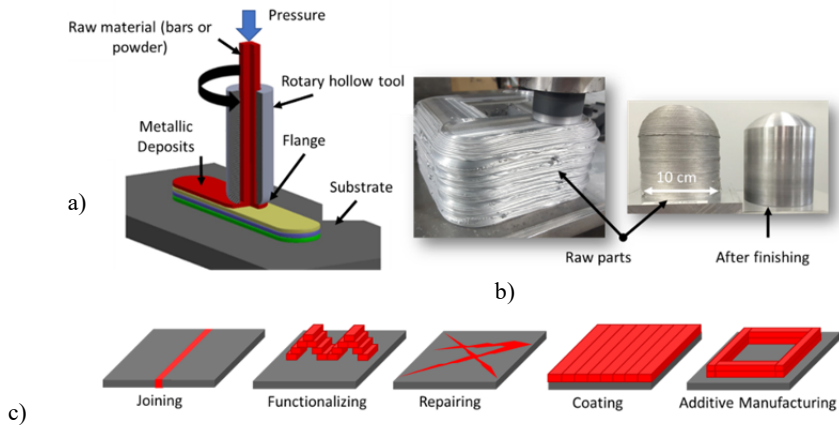


Figure 1.72. a) Schematic representation of the MELD solid-state material deposition process. b) Rough aluminum MELD part before and after mechanical finishing. c) Types of applications: the MELD process allows you to build, repair, functionalize and coat (MELD 2019)

The process makes it possible to build a whole part, coat a surface, add a new function, make a repair or join or even combine all of these operations. In addition, the material deposition takes place in an open atmosphere and allows the manufacture of large parts. It is also very fast in terms of the deposition rate compared to L-PBF or E-PBF technologies. Thus, the manufacture speed declared by MELD Manufacturing is $1020 \text{ cm}^3 \cdot \text{h}^{-1}$ for an aluminum alloy ($622 \text{ cm}^3 \cdot \text{h}^{-1}$ for steel) versus only 50 and $80 \text{ cm}^3 \cdot \text{h}^{-1}$ for L-PBF and E-PBF, respectively.

In addition, the associated thermomechanical effect makes it possible to obtain uniform fine-grained microstructures, which contributes to improving the properties of the parts produced (elastic limit, fatigue resistance) (MELD 2019). The technology is also promising in terms of the development of new materials. It is compatible with alloys of titanium, aluminum, copper, nickel-based superalloys (Inconel) and steels (316L). It also opens up prospects for numerous studies, in particular for composites with a metal matrix based on aluminum or copper (Griffiths et al. 2019), or even for obtaining property gradients or multi-materials. On the other hand, the building precision is around a centimeter, which systematically requires significant mechanical finishing.

1.6. Conclusion

This chapter has presented a general state of the art on the most currently used metal additive manufacturing processes, starting with the processes involving direct material melting (sections 1.1–1.4). Section 1.5 gave a brief overview of the new processes under development, starting with indirect processes with solid-phase laser sintering and ending with processes involving plastic deformation (MELD, cold spraying). A description of the physical mechanisms involved in metal fusion AM processes (DED, L-PBF, E-PBF, WAAM) is given in Chapter 3 of this volume.

A summary of the main characteristics of the four main metal fusion additive manufacturing processes is presented in Table 1.4, as well as a (non-exhaustive) list of machine manufacturers in Table 1.5. The performance of these machines is constantly improving, both in terms of robustness/reliability (gas flow management in L-PBF machines) and in terms of the dimensions of parts that can be produced and productivity. Thus, the German company SLM Solutions GmbH released, at the end of 2020, its new powder bed melting machine (dimensions $600 \text{ mm} \times 600 \text{ mm} \times 600 \text{ mm}$), equipped with no less than 12 lasers of 1 kW, with productivity increased by more than a factor of 10 compared to previous generations of machines (SLM Solutions 2020).

Characteristics	DED	L-PBF	E-PBF	WAAM
Power used (W)	300–2,000	100–500	1,000–5,000	1,000–6,000
Scan velocity V_0 (mm·s ⁻¹)	2.5–25	500–1,500	1,000–6,000	5–10
Layer height Δh (mm)	0.1–1	0.02–0.1	0.03–0.15	0.2–2
Energy source diameter D_0 (mm)	0.5–3	0.05–0.15	0.1–0.2	1–4
Theoretical maximum hourly flow rate (cm ³ ·h ⁻¹)	~250	~50	~600	~300

Table 1.4. Comparative characteristics of metal AM processes. The values of the maximum hourly flow rate, estimated as $\Delta h \times D_0 \times V_0$, are related to an energy source and can increase with the number of sources integrated in the machines (L-PBF machines with four lasers)

Type of process	Manufacturers
DED-LMD	BeAM – ADDUP (France) ⁴ Optomec (USA) ⁵ ; Sciaky (USA, EBAM technology) ⁶
L-PBF	SLM Solutions Gmbh (Germany) ⁷ ADDUP (France) ⁸ EOS (United Kingdom) ⁹ Renishaw (United Kingdom) ¹⁰ 3D-Systems (USA) ¹¹ ; Concept-Laser (GE Additive, USA) ¹²
E-PBF	Arcam (GE, USA) ¹³
WAAM	FRONIUS ¹⁴

Table 1.5. Non-exhaustive list of industrial machine manufacturers

4 Available at: <https://www.beam-machines.fr/>.

5 Available at: <https://optomec.com/3d-printed-metals/lens-technology/>.

6 Available at: <https://www.sciaky.com/fr/fabrication-additive/technologie-de-fabrication-additive-par-faisceau-d-electrons>.

7 Available at: <https://www.slm-solutions.com/>.

8 Available at: <https://www.addupsolutions.com/>.

9 Available at: <https://www.eos.info/en/additive-manufacturing/3d-printing-metal>.

10 Available at: <https://www.renishaw.com/en/metal-3d-printing-32084>.

11 Available at: https://www.3dz.fr/product_manufacture/3d-systems/.

12 Available at: <https://www.ge.com/additive/who-we-are/concept-laser>.

13 Available at: <https://www.ge.com/additive/additive-manufacturing/machines/ebm-machines/arcam-ebm-spectra-h>.

14 Available at: <https://www.fronius.com/en/welding-technology/info-centre/magazine/2019/waam>.

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2

Raw Materials: Metal Powders and Wires

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The raw material, whether powder or wire, is, in the same way as the energy source, the main ingredient for manufacturing parts with additive processes. Good quality control of this raw material is therefore essential for the overall process management. This chapter provides an overview of the main factors concerning powders and metallic wires.

2.1. Metal powders

2.1.1. Introduction

The quality requirements for industrial parts produced by additive manufacturing (AM), but also the safety requirements in sectors such as aeronautics or nuclear power, demand strict control across the entire value chain of the feedstock. The first stage concerns the use of a raw material that is ideally suited for the concerned technology and thus to its specific characteristics. Indeed, each machine is characterized by its specificities and its own tolerances, which means that the powders must meet criteria adapted to the targeted technology and even to the manufacturer for a given technology.

Section 2.1 written by Marc THOMAS, Stefan DRAWIN and Olivier BONNEFOY.

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The high quality of the raw material (powder) certified by the supplier is a necessary, but not sufficient, condition to guarantee the manufacture of parts without defects or with some controlled defects for reproducible properties. The powder can thus be a source of inconstancies and irregularities during successive builds and, in a deceptive manner, the resulting mechanical properties of materials can be altered and may vary over time. However, other factors such as the laser process conditions may also be responsible for poor part quality, but these are not the focus of this chapter. Thus, an extended characterization of the powders, their systematic control and their increased traceability during their lifecycle are essential steps for the qualification of AM parts.

This chapter is therefore devoted to metal powders, providing a description of this raw material in relation to the specific needs of the various AM processes. The important requirements in terms of raw material quality have resulted in a standardization effort, in particular for characterization methods such as the relatively recent ASTM F3049-21 Standard Guide for Characterizing Properties of Metal Powders Used for Additive Manufacturing Processes, published jointly by the international standards organizations ASTM and ISO (ASTM 2014). The main technological challenges related to this feedstock can be summarized as follows:

- a lack of fine powder characterization methods (chemistry, structure), which correspond to important input data;
- a lack of clear relationships established between the intrinsic properties of the powders and the final properties of the produced materials;
- a lack of understanding of the factors underlying the problems of variability from one machine to another or from one day to another for the same machine, which may be related to the reproducibility of the powder or to some machine wear (for example, the layering system);
- the lack of simulation tools for powder manufacturing processes (atomization, etc.) that could be used for optimization;
- the inadequacy of referring to traditional methods of qualification and certification of materials for the qualification of parts;
- the still too limited qualification of metal powders, which makes it impossible, for example, to clearly identify the differences between a new powder and a recycled powder;
- the still too fragmentary nature of material databases for AM with precise control and reliable traceability.

2.1.2. Producing powders

The metal powders used for AM are, originally, essentially derived from powders commercially available in the powder metallurgy sector, which are intended to be sintered in the solid phase or extruded, but in no case intended to be melted. The problems are, however, similar to the case of the L-PBF process: we want a mold (or a powder bed of a given thickness) to contain a minimum of void between the particles, so that the shrinkage during consolidation by sintering (or during laser or electron beam melting) is as low as possible and the residual porosity is minimized. This imposes constraints on the powder characteristics, such as the particle size, their size distribution, their morphology and the overall rheological behavior of the powder. The processes providing powders with “good” characteristics have thus been favored. However, producing “ideal”, or almost ideal, powders is expensive. Therefore, a compromise between the quality of the powder and its price must be found for each particular case, depending on the properties we are looking for in the parts produced by AM. It is also important to bear in mind that the powder quality is only one of the factors that influences the quality of the parts, in addition to the choice of the AM process, the operating parameters used, the post-treatments selected, etc.

First, for simplicity, we assume that we are looking to obtain spherical (capable of more compact arrangements than shapeless particles) and smooth (providing more fluid powders than rough particles) particles.

Atomization is among the processes most commonly used to produce powders for AM, consisting of fragmenting a liquid jet (here a liquid metal) with another fluid which, at a high flow rate, shears the liquid jet and transforms it into droplets, in one or more stages: this then introduces the notion of *atomization with an inert gas, air or water*.

In the metal–fluid interaction zone, the metal can also be introduced in solid form (e.g. a wire), then melted and atomized simultaneously. To do this, the atomization fluid must have sufficient temperature and kinetic energy: for this, plasmas are used, generated either by an electric arc in a nozzle, a radiofrequency field or microwaves. This is called plasma atomization.

The installation is generally referred to as an *atomizer* or *atomization tower*.

2.1.2.1. Gas atomization

The principle of gas atomization is shown in Figure 2.1. The melting of the metal takes place at the top of the facility. For a small volume of metal (a few liters),

melting is carried out by inductive heating in a ceramic crucible (usually alumina), such as in a vacuum induction melting (VIM) furnace; the process is thus known as vacuum induction melting inert gas atomization (VIGA). The crucible has a channel in its lower part, the orifice of which is closed by a vertical bar (stopper rod, not shown in the figure), during the heating phase; the bar is then raised to allow the liquid metal to flow.

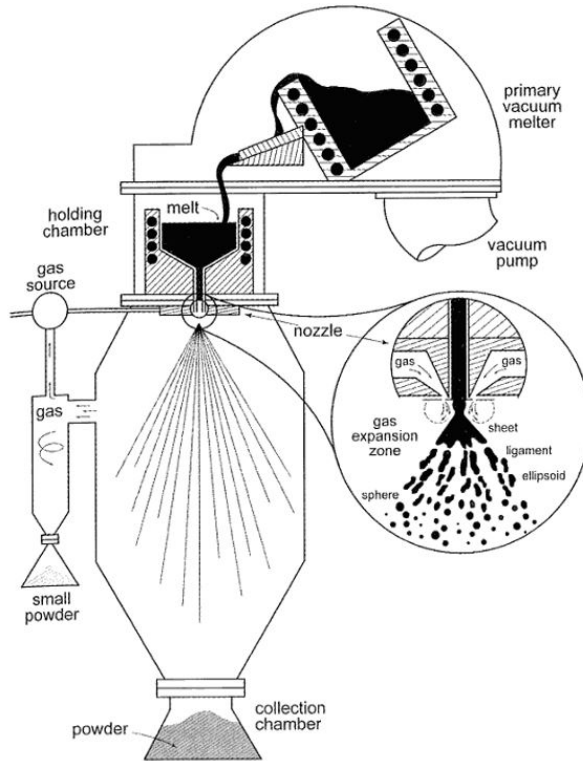


Figure 2.1. *Principle of gas atomization according to German (2005)*

For larger volumes, more than 200 L (1.6 tons of steel, for example), the melting takes place in an additional module; the melt is gradually poured from a primary furnace into an intermediate tank, which feeds, via a channel, the interaction zone with the atomization gas. The dissociation of the large-volume melting furnace/small supply tank makes it possible, because of simple safeguards (e.g. that stop the primary furnace tilting), to limit the risks of atomizer deterioration, which could cause an accident in the furnace containing several hundred kilograms of molten metal placed directly above the installation. In addition, the level of liquid

metal in the tank is practically constant during most of the atomization cycle, which guarantees a constant flow of liquid metal during this period.

The atomization gas is injected at high pressure and flow rate (20–60 bar, 15–40 m³/min, depending on the size of the installation) into an annular nozzle opening at the outlet of the liquid metal supply channel (see insert in Figure 2.1).

The actual atomization occurs in the *atomization chamber* at a pressure close to atmospheric pressure. Metal droplets, resulting from the sheets and ligaments created by the shearing of the liquid metal by the gas (see insert in Figure 2.1), are formed in the metal/gas interaction zone. They are shapeless at the beginning, but under the action of the liquid/gas surface tension, they spheroidize during their drop, while they are still liquid, and finally solidify and reach the walls or the bottom of the chamber, which are relatively cold.

There are then two operations to perform: collecting the powder and emptying the atomization gas. Several configurations are possible.

In Figure 2.1, the heaviest (largest) particles fall by gravity into a container at the bottom of the installation. The gas, carrying the lightest (finest) particles, exits through a side duct connected to a cyclone, which has the function of separating the particles from the gas.

For small installations, in particular those for R&D, the design is more compact: the gas carries all of the particles to the cyclone. The particles, except the lightest ones, then flow into a container connected to the cyclone.

In all cases, the gas passes through a filtration unit to trap the very fine particles that were not captured by the cyclone.

Finally, the gas is ejected into the atmosphere or otherwise directed to a recycling unit.

This basic principle for gas atomization has many variations. Three examples are provided as follows:

- the melting chamber can be pressurized by a gas, which then exerts pressure on the liquid metal in the casting ladle and which has the effect of increasing its mass flow. This of course requires a good gas tightness between the melting chamber and the atomization chamber;
- most often, the gas is ejected from the atomizing nozzle into an annular part coaxial with the flow of the liquid metal. The gas outlet section, instead of being

continuous, can be made up of multiple orifices (up to a dozen), which create many gas jets converging on the liquid metal flow. This multi-jet configuration can also be achieved with independent gas inlets, by injectors arranged axisymmetrically a few centimeters below the liquid metal outlet and oriented toward the flow (discrete multi-jet configuration);

- for the production of aluminum powders, reverse atomizers are often used. The injection of liquid metal and gas takes place at the bottom of the chamber: the nozzle is in contact with the upper surface of the melt-pool and the liquid is carried by the vacuum created by the atomization gas at the orifice of the outlet (suction effect). The gas leaving the cyclone is reinjected along the internal wall of the atomization chamber. This configuration has three advantages:

- there are fewer gas recirculation loops in the chamber, in particular in its lower part, and thus fewer collisions between the particles carried in these loops and therefore fewer satellites (small particles clinging to large particles),

- the particles are diluted by this gas supply, which improves process safety (reduced risk of explosion of this pyrophoric particle cloud),

- the parietal gas flow reduces the quantity of particles sticking to the chamber wall, which also improves the process safety (reduced risk of ignition of this layer of fine particles when the liner is ventilated).

In this configuration, the gas and liquid metal flow rates are no longer independent: the design of the nozzle is therefore of great importance to generate optimal atomization conditions.

2.1.2.1.1. Atomizing gas

The choice of the atomizing gas affects the properties of the powder and the process cost. Three types of gas are generally used, from the most expensive to the cheapest: noble gases (most often argon and sometimes helium), nitrogen and air. As the latter two can react with some alloys in the liquid state (nitriding, oxidation), the question must be asked for each alloy whether the rate of nitrogen or oxygen contamination is compatible with the intended uses for the powder and whether a low-cost purification route is available.

It is important that air and nitrogen only affect the particle surface, which means that the high-temperature solubility of these two species in the considered alloys should be low. Fine particles are more contaminated with O and N than coarse particles because of their larger specific surface area (expressed in m^2/g). An example of a deoxidation treatment consists of heating the powder in a fluidized bed in an atmosphere containing a small amount of hydrogen.

For some fine alloy powders such as Al and Mg, which react violently with the air during various manipulations (sampling, sieving, packaging, etc.), a passivation treatment greatly reduces the risks of ignition and explosion. This consists of adding approximately 0.5% by volume of O₂ to the atomization gas (Ar, He, N₂). Using air, a dense oxide layer (alumina) ~5 nm thick forms naturally, in which is sufficient to prevent incidents. However, the air atomization of aluminum alloys remains dangerous: the volume density of the powder particles in the chamber must remain low, below the flammability limit.

The viscosities of Ar, He, N₂ and air gases are similar, while other physical properties are different (density, thermal conductivity, heat capacity, etc.). Through the same injection nozzle and for the same atomization pressure (upstream of the nozzle), the helium jet will have a mass flow approximately three times lower than argon but a gas velocity approximately three times higher. Some investigations (Ünal 1990; Özbilen 2013) carried out for aluminum and magnesium alloys have shown that ligaments can be more efficiently disrupted by this higher speed, thus generating finer particles than in argon. In nitrogen, of intermediate density between Ar and He, an intermediate median diameter is also obtained. The presence of oxygen in the atomization gas modifies the surface tension of the ligaments and the droplets through the formation of a low ductility oxide layer, so that their spheroidization is incomplete: this therefore generates irregularly shaped particles.

2.1.2.1.2. Inert gas atomization of reactive or refractory alloys

Some alloys are not compatible with conventional alumina crucibles, either because of their melting temperature being beyond the capabilities of alumina or because of their composition. This is especially the case for alloys with a high content of titanium (Ti, TiAl, etc.), zirconium or niobium, which react chemically at high temperature with alumina (e.g. $\text{Al}_2\text{O}_3 + \text{Ti} \rightarrow \text{Al}_2\text{O}_{3-x} + \text{TiO}_x$), which pollutes the liquid metal and may lead to crucible fracture. Two solutions are commonly applied.

Use of a cold crucible: A cold crucible consists of a set of segments cooled individually by circulating water (Figure 2.2). The metal load is heated by the circulation of eddy currents created by a spiral inductor external to the crucible. The frequency of the magnetic field and the injected electric power must be adjusted to the volume of the alloy to be melted (and therefore to its mass), so that the load is semi-levitated in the crucible thanks to the Lorentz forces. The latter detach the molten alloy from the side walls of the crucible, greatly reducing heat exchange between the hot alloy and the cold crucible and thus eliminating the risk of chemical

reaction. When melting a large amount of alloy (a few liters (4–15 kg of titanium) to a hundred liters (400–500 kg of titanium)), the Lorentz forces are insufficient to levitate the overall alloy. In these cases, some of the alloy solidifies on contact with the crucible and creates a skull in which the majority of the alloy remains liquid. This solution requires a very high heating power, most of which is absorbed by the water cooling circuit. To melt 5 kg of titanium alloy, around 250 kW is required, while for 400 kg, more than 2 MW is needed.

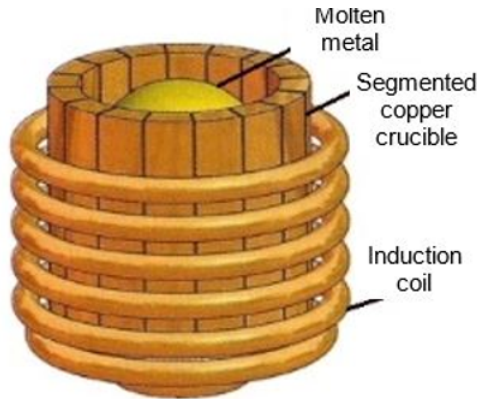


Figure 2.2. Cold crucible according to Metal Powder and Process Ltd

Electrode induction melting inert gas atomization (EIGA): This crucible-free process is mainly an adaptation of the alloy melting method. It was developed in the 1980s specifically for titanium alloys, but can be used for other materials, reactive or not, up to melting temperatures of over 2,400°C. A bar of the material to be atomized, with a conical tip at its lower end, is suspended above a conical inductor (Figure 2.3). The surface of the tip is melted, creating a stream of liquid metal that flows by gravity through the atomizing nozzle into the metal–gas interaction zone. As the bar is consumed, it is lowered at a speed set by the operator, so that the tip remains in a fixed position relative to the inductor, thus ensuring inductive coupling and a constant liquid metal flow. The bars have a diameter of about 50–150 mm, for lengths of between 500 and 1,000 mm; the geometry of the inductor must be adjusted to the diameter of the bar. For this contactless EIGA process, the nozzle configuration is called “free-fall” while it is “close-coupled” for the VIGA process (Figure 2.4).

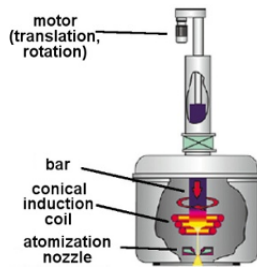


Figure 2.3. EIGA melting module according to ALD-VT¹

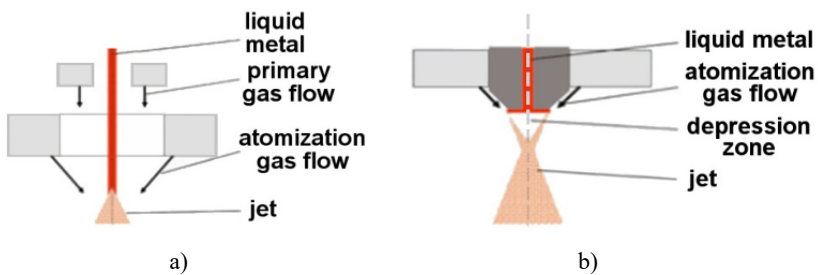


Figure 2.4. Configuration of atomization nozzles according to Indutherm (2020): a) free-fall; b) close-coupled

2.1.2.2. Water atomization

Water can replace gas as an atomizing fluid. The advantages are the low cost of the fluid, its high density (a thousand times greater than that of air), its incompressibility and the possibility of using high pressures (1,500 bar and more) and therefore high jet flow rates (500 m/s and more) if the production of fine particles is desired. The disadvantages are as follows: (i) the high cooling rate for the molten metal, producing particles of irregular shape (spheroidization not taking place), with a rough surface (Figure 2.5) and high porosity (hence often called “sponge”); (ii) the oxidation of the particles at the surface (for oxidation-sensitive alloys), which can be limited because the time spent at high temperature is very short; and (iii) the possible production of hydrogen upon liquid metal–water contact. Added to this is the need to dry the powders and possibly deoxidize them.

Under certain atomization conditions (high supercooling, high atomization pressure; the metal is still very fluid and finer particles are produced), the

¹ <https://www.ald-vt.com>.

characteristic spheroidization time becomes smaller than that of cooling, so that fine particles ($<15\text{ }\mu\text{m}$) become almost spherical, as in the case of aluminum alloys.

The powder purity and the process safety can be improved by the addition of oxidation inhibitors to the water, the use of deionized water or the addition of alcohol.

Oils can be used in place of water if a reduced contamination with oxygen is desired. This is not possible if the presence of carbides formed in contact with oils deteriorates the properties of the alloys.

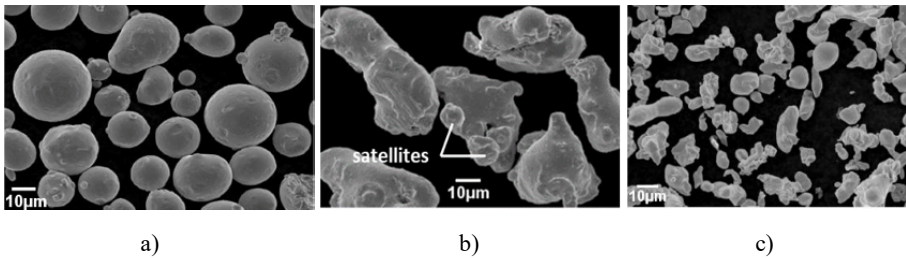


Figure 2.5. SEM micrographs of 316L steel powders produced by (a) gas and (b, c) water (two atomization conditions) atomization, according to Hoeges et al. (2017)

2.1.2.3. Plasma atomization

Plasmas are a source of heat, capable of melting the most refractory materials. They are used for the production of powders from wires. These processes are well suited to sufficiently ductile materials for which wire drawing is possible, and the wire is compatible with reels and pulleys with a small radius of curvature.

Plasma torch atomization: The wire, with a diameter less than 6 mm, is unwound in a chamber placed under reduced pressure ($<1\text{ atm}$), at the apex of two or three torches (Figures 2.6(a) and 2.7) in which an electric arc creates a plasma in an argon-based gas. Thanks to their high speed, the plasma jets melt the wire and eject the liquid material in the form of droplets, which spheroidize and solidify during their fall. Due to the reduced pressure, the low gas recirculations in the chamber, the high supercooling generated by the high temperature of the plasma jets and the relatively low metal mass flow rate ($<10\text{ kg/h}$), few sheets and ligaments are formed and few collisions take place between droplets: the particles are rather fine, with a well-defined spherical shape, only a few satellites and no internal porosity (Figure 2.6(b)).

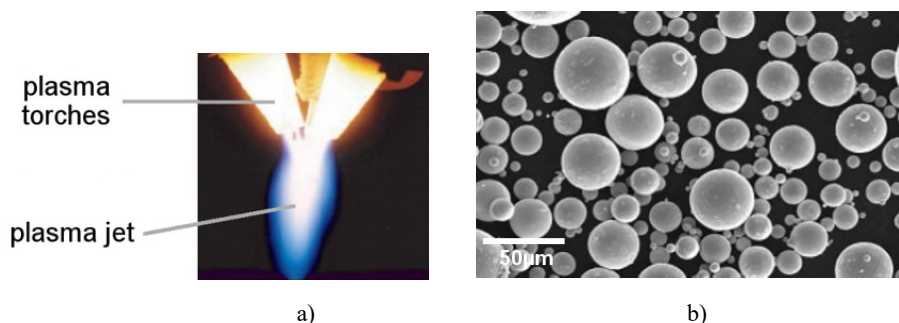


Figure 2.6. Plasma atomization: a) plasma torches; b) plasma-atomized powders (pyrogenesis)

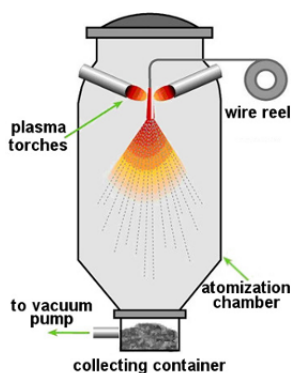


Figure 2.7. Principle of plasma atomization (AP&C)

The process has been improved over time to increase the powder mass flow rate, the energy efficiency and the fine particle fraction ($<45\text{ }\mu\text{m}$, consistent with AM) and reduce the overall powder cost by adding a wire preheater, increasing the gas injection pressure and changing the plasma/wire interaction zone (which takes place in a prechamber, with liquid and gas being expelled into the main chamber via a De Laval nozzle). This process is particularly suited to titanium and its alloys (inert atmosphere, absence of crucible), but not to intermetallics (e.g. TiAl), which are not ductile enough.

Induction plasma atomization: The wire, positioned on the axis of the device (Figure 2.8), is gradually heated by the plasma until it reaches its melting temperature (or even higher) above the atomization nozzle. The plasma drives the

liquid through the nozzle; at its outlet, the gas (plasma) expansion leads to the formation of droplets. The droplets cool as they fall into the chamber located under the nozzle (not shown here).

In conclusion, the previously illustrated processes are the main ones for mass production of the alloy families most used to date for AM. One notable difference is the shape of the raw material (bulk, bar, wire, powder). However, the characteristics of the resultant powders depend on the atomization process; this point is dealt with in section 2.1.3.

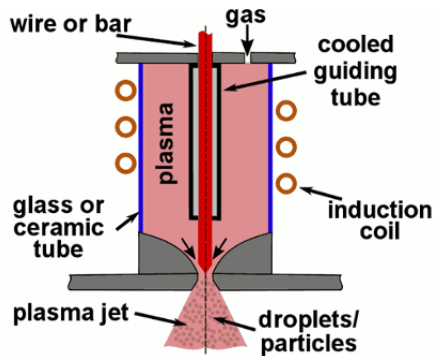


Figure 2.8. Principle of induction plasma atomization. For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

2.1.2.4. Other methods of powder production – spheroidization

There are many techniques for producing metal powders, but not all are suitable for AM specifications and/or on an industrial scale.

The (plasma) rotating electrode process (PREP or REP) provides excellent quality powders for parts to be densified by hot isostatic pressing (HIP), exhibiting a well-defined spherical shape with few satellites and without internal porosities. However, the output is low: a bar that is rotated on its axis at a speed greater than 10,000 rpm must be machined with high precision so as to be accurately balanced; this imposes limits on its size and powders are therefore only produced in small batches. In addition, the particles obtained are generally too large ($d_{50} > 100\text{--}150\text{ }\mu\text{m}$) for this process to be suitable for AM (this would need the sieving out of the finest fraction) and economically viable.

For pure titanium, chemical (using chlorine, magnesium, sodium, hydrogen, etc.) or electrolytic processes can convert TiO_2 , extracted from the ore and purified, into

Ti metal in powder form after final grinding (Fray/FFC/Cambridge, Armstrong, CSIRO, MER, HDH or hydride-dehydride processes, etc.) (Froes 2012). The production of alloys is possible, although it requires good control of the reactivity of the added elements, in particular for the Ti64 alloy (Ti-6Al-4V, in percent by mass). However, these powders are angular, sometimes flat or elongated, and not at all spherical (Figure 2.9). Almost all these processes, less expensive than gas or plasma atomization, are still under development.

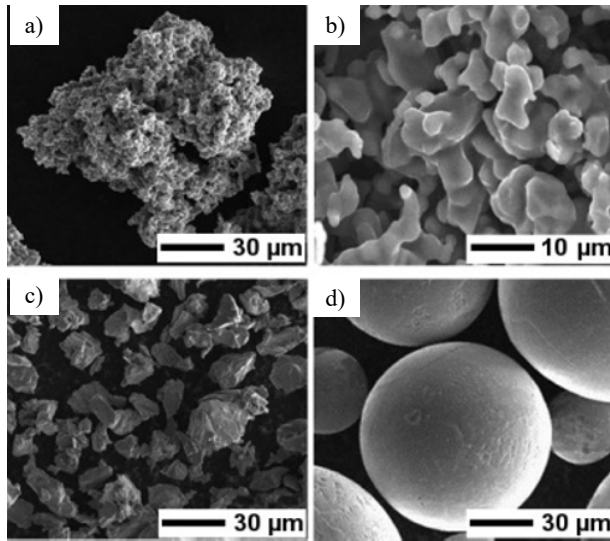


Figure 2.9. SEM micrographs of Ti64 powders produced via different processes: a) Armstrong, b) FFC, c) HDH and d) PREP. From Fang et al. (2018)

Pre-alloyed or *pre-diffused* powders can be produced by mechanical alloying (through high energy grinding in a planetary grinder or an attritor), for example when nano-structured particles are pursued or even amorphous ones, or for the introduction of a uniform dispersion of hardening nano- or micro-particles into an alloy (for example ODS steels). The resultant particles are angular and therefore, *a priori*, not very suitable for AM.

These shapeless powders, as well as those resulting from water atomization, can however be *spheroidized* by passing through a radiofrequency plasma (Park et al. 2020). The particles are melted, becoming spherical thanks to the surface tension, and are collected at the outlet: they are then well-adapted to AM. The injection of

very pure gases (argon, helium) into the chamber leads to a reducing atmosphere, so this post-process decreases the oxygen content of the particles, which is beneficial for many alloys. However, the major composition elements of some alloys can be slightly modified, as some elements (Cr, Al, etc.) tend to evaporate while the particles are in the liquid state.

Everything that has been tested in laboratories is impressive, ultimately giving fairly good preliminary results in AM: non-spherical powders (such as those described above) (Vasquez et al. 2017), mixtures of powders with very different median diameters (d_{50}), mixtures of elementary powders (Polozov et al. 2018), etc. However, not all these studies will reach industrial maturity, because the final characteristics of the resultant parts (porosity, microstructure, mechanical behavior, etc.) must meet high criteria in terms of reproducibility.

2.1.2.5. Influence of atomization parameters

For gas atomization, equation [2.1], established empirically by Lubanska (1970), shows the influence of some operating parameters on the particle size, described simply by the median diameter, for a given atomization geometry:

$$d_{50} \sim K \cdot D \cdot \left(\frac{v_m}{v_g We} \left(1 + \frac{Q_m}{Q_g} \right) \right)^{1/2} \quad [2.1]$$

where K is a constant, D is the diameter of the melt flow, Q_m is the metal mass flow rate ($\text{kg} \cdot \text{s}^{-1}$), Q_g is the gas mass flow rate, $v_{m,g}$ is the kinematic viscosity ($\text{m}^2 \cdot \text{s}^{-1}$) of the metal and gas (respectively) and We is the dimensionless Weber number, $We = \frac{\rho_m u_g^2 D}{\gamma_m}$, where ρ_m is the density of the melt ($\text{g} \cdot \text{cm}^{-3}$), u_g is the speed of the gas on impact with the liquid metal ($\text{m} \cdot \text{s}^{-1}$) and γ_m is the surface tension of the liquid metal ($\text{N} \cdot \text{m}^{-1}$).

In order to decrease the particle size, the ratio Q_m/Q_g must be reduced by using a high gas flow rate (at a given metal flow rate), and the Weber number must be increased by increasing the gas velocity. The latter can be achieved by reducing the outlet area of the gas injection nozzle and using high gas pressures to maintain a high flow rate. In VIGA atomization, the diameter D is related to the area of the outlet channel in the crucible: decreasing D certainly decreases d_{50} , but also reduces Q_m (unless gas pressure is applied above the crucible to force the liquid to flow), which increases atomization time, crucible wear and gas consumption and may make it more difficult for metal to flow, or even create a solidified metal plug if the duct is

not hot enough. In EIGA atomization, the diameter D is not controlled and mainly depends on ρ_m , v_m and γ_m .

Raising the temperature of the metal, by heating it beyond its melting temperature in the crucible (supercooling), reduces its viscosity v_m and its surface tension γ_m , which helps to reduce the size of the particles (the sheets and ligaments in the liquid are less resistant to the shear imposed by the gas flow). Raising the temperature of the gas increases its kinematic viscosity v_m and helps reduce the d_{50} .

The atomization geometry is also very important: the type (conical annular nozzle, multi-jet, etc.), geometry (gas injection angle, adjustment of the gas and melt outlet area) and dimension (distance between nozzle and point of impact with the melt) of the injection nozzle define the efficiency of the mechanical interaction between gas and melt. Moreover, all the elements (gravity flow axis of the melt, point of convergence of the gas jet(s) and its/their intersection with the axis of the liquid flow, etc.) must be correctly aligned, and this throughout the atomization cycle.

In practice, once the atomization geometry has been set, a supercooling temperature for the liquid metal (in VIGA) is defined, and the gas flow rate is adjusted via the pressure at the atomizer inlet. In EIGA, the temperature of the metal is difficult to control, as liquid can flow down before the metal is even completely liquid; the metal flow is regulated by the electrical power injected into the inductor, and the gas flow via the gas pressure.

2.1.2.6. *Functionalization of powders*

Gas or water atomization processes do not always make it possible to obtain powders suitable for AM. Indeed, the presence of non-spherical particles and “satellites” can contribute to poor powder flowability or spreadability. It may thus be advantageous to subject the non-“optimal” powder to a post-treatment which involves improving its surface properties. There are two kinds of processes that can achieve this result.

The first kind is intrusive since it consists of a liquid treatment used to obtain a perfect spheroidization of the powder. It is explained in section 2.1.2.3. Several manufacturers have specialized in the production of quasi-spherical powders by developing innovative processes:

- the TEKNA process with induction coupled plasma (ICP) atomization: this process uses a solid precursor (sponge, powder, etc.) and an ICP plasma for remelting and spheroidization;

- the LPW Technology Ltd and Metalysis process: this association combines the production of Metalysis powders and the LPW spheroidization technique. The production of powders at Metalysis is obtained through the electrolysis of an oxide. This is an economical process (FFC Cambridge), which works by introducing metal oxides into a bath of molten salt (calcium chloride);

- the H.C. Starck and Metaspheer Technology process: this is another partnership between a powder manufacturer and a powder processor. The Metaspheer process makes it possible to restructure electrically conductive materials on a nanometric scale to spheroidize them.

The second way to functionalize the powders involves forming a deposit on the surface of the particles either to improve the morphology of the powder or to modify its chemistry. Surface modification by coating can be obtained by the fluidized bed technique and can include one or more layers (LIFCO 2020). The surface-functionalized powders are generally multi-material (metal/metal, ceramic/metal, metal/polymer) to meet multifunctional specifications. There are many traditional fields of application: anticorrosion coating on Fe powder, metallic coating with high electrical conductivity on ceramic powders, coating with tribological properties, etc. It could indeed be interesting to improve the flowability of the powder through this type of functionalization, but still taking into account the resultant chemical modifications (see section 2.1.4.4 on strategies to improve the rheology).

2.1.2.7. Powder suppliers

The largest powder suppliers generally correspond to the companies that atomize the powders, such as the Eramet group in France. Via its subsidiary Erasteel, Eramet produces gas-atomized metal powders of various steels and nickel-based superalloys using two industrial atomization towers of 10 and 14 tons. Due to the growing AM market, these companies are adapting to customer needs, for example, by favoring a given particle size depending on the type of AM machines. On the other hand, it is possible to also obtain supplies from powder resellers who can offer batches of standard powders at competitive prices. Finally, a third solution exists, which is to purchase the powder directly from the AM machine manufacturer, often with the guarantee of being able to benefit from the correct parameters for the powder purchased, but this is also the most expensive solution.

As a reference, a non-exhaustive list of powder suppliers is presented in Table 2.1 with their specificities and corresponding websites.

AMETEK https://www.ametekmetals.com	METALPINE GmbH http://www.metalpine.at
A.M.P.E.R.E Alloys https://www.amperemetalpowders.com	Metal Powder and Process Ltd https://metalpowderprocess.co.uk
AP&C http://advancedpowders.com	Mimete metal powders http://www.mimete.com
ASM https://asm-au.com	NanoAL http://www.nanoal.com
ERAMET http://www.eramet.com	OSAKA Titanium technologies Co. Ltd. http://www.osaka-ti.co.jp/e
CARPENTER https://www.carttech.com/en	OERLIKON AM https://www.oerlikon.com/am/en/
EQUISPHERES https://equispheres.com	PRAXAIR Surface Technologies http://www.praxairsurfacetechnologies.com
GKN Powder Metallurgy https://www.gknpm.com	PYROGENESIS Canada Inc http://www.pyrogenesis.com/fr/produits-services/poudre
HÖGANÄS AB http://www.hoganas.com	SAFINA http://www.safina.ro/en/products/atomized-powders
KOLIBRI Metals GmbH http://www.kolibri.de.com	SANDVIK https://www.materials.sandvik/fr
KYMERA International https://www.kymerainternational.com	TEKNA http://www.tekna.com/fr
LIBERTY POWDER METALS http://www.libertysteelgroup.com	TLS-Technik http://www.tls-technik.de
LPW TECHNOLOGY Ltd https://www.lpwtechnology.com/fr	VDM Metals International GmbH https://www.vdm-metals.com

Table 2.1. *List of powder suppliers (non-exhaustive)*

2.1.2.8. Powders developed for AM

There is a strong demand from users of metal powders for the best possible quality in terms of traceability, storage and delivery, but also reproducibility from one batch of powders to another. For example, it is considered beneficial to use rather narrow particle size distributions to avoid segregation during transportation and handling. On the other hand, in the case of “powder bed” processes, for the powder to be spread efficiently in a bed with a homogeneous thickness and high density, it is necessary to favor a wide particle size distribution so that small ones fit between the large ones. Another challenge is to keep the particles in an unmelted state and not partially melted, as this may cause the formation of cavities, which are

difficult to detect. In the case of LMD processes, it is necessary that the powders are transported with a constant mass flow rate, without risk of nozzle clogging (total, partial or temporary) by agglomerates.

This recent development has led many powder suppliers to adapt their “products” for a label that meets known criteria for AM. Here, are a few examples:

- Addalloy (5T, 7S, HX), aluminum alloys from NanoAL;
- AMPERPRINT from H.C. Starck;
- AncorTi (titanium) from Höganäs AB;
- BLDRmetal from Formetrix/Nanosteel;
- Böhler AMPO (steels and nickel alloys) from the Voestalpine group;
- MetcoClad 718, MetcoClad 625 and MetcoClad 625F from Oerlikon;
- MS1 (Maraging 300) from EOS;
- Pearl Micro from Erasteel;
- TILOP (Titanium Low Oxygen Powder) from Osaka;
- TruForm from Praxair Surface Technologies;
- VDM Alloy 699 XA and VDM Alloy 718 CTP from VDM Metals.

Even when all the challenges have been overcome, the mechanical tests carried out on the finished products often reveal altered properties compared to the equivalent material obtained by the conventional route. It should be remembered that AM processes involve extremely rapid melting and solidification, which gives the resultant materials very specific characteristics (nanostructuring, low precipitate density with supersaturation of certain elements, high density of dislocations). These characteristics are generally beneficial for the properties of the alloys under consideration, in particular with increased strength, but this can also induce detrimental metal brittleness. This is the reason why it is necessary to establish the relationships between the microstructure and the texture obtained on products after post-treatments and their mechanical properties. Many teams of researchers are now looking for the optimum, not only through thermal post-treatments but also through a change in the chemistry of the alloys. The grades developed for conventional processes are not necessarily suitable for AM processes. There are numerous recent developments in the field of post-heat treatments in particular and the finishing stage in general, but also in the field of alloy composition modification or even the exploration of new alloy systems.

2.1.3. Physico-chemical properties of powders

The characterization of powders aims at establishing a certain number of quality criteria for the powder intended for AM. Overly large variations in the characteristics of the powder can indeed lead to non-uniform layering and a non-homogeneous bulk density in the powder bed, which can lead to a poor surface condition for the part and to defects below the surface, resulting in undesirable mechanical properties. This explains the importance of identifying the most influential powder characteristics to ensure the expected and reliable performance level of the manufactured parts.

There are a number of so-called optimal powder characteristics: a narrow particle size distribution with a spherical morphology for good flowability in DED (but a bimodal size distribution of particles in L-PBF for a good compacted layer), a low level of surface contamination and a reduced volume fraction of intraparticle pores.

Conversely, a number of non-optimal powder characteristics can be listed: batch-to-batch inconsistency in terms of variations in chemical composition, inadequate particle size distribution, imperfect sphericity of particles (elongated, angular, flat), the presence of “satellites” on the surface of large particles, the development of an oxide film or the presence of inclusions on the surface of particles, the preservation of intraparticle porosities (pores of occluded gas) and sensitivity to interstitial contamination (O, N, C, etc.).

The physico-chemical properties of powders are important to know in order to guarantee the quality of the finished product and to ensure higher productivity. Different experimental techniques can thus be used, for example, measurement of the particle size distribution by sieving and by laser particle size distribution (dry or wet) and chemical analysis by inductively coupled plasma–optical emission spectroscopy (ICP-OES). It is also important to investigate the morphology of particles under a scanning electron microscope (SEM) associated with image analysis for quantification purposes, the porosity ratio using an SEM and He pycnometer, measurements of densities (poured and tapped) and flowability. Several methods intended to evaluate these physico-chemical properties have now been identified.

– *Particle size distribution*: This corresponds to the “percentage by mass, by number or by volume, of each fraction into which a powder sample has been classified with respect to size” according to ISO 3252:1999 (en). Sieve analysis (typically 40-63-80-100-125-160-180-200-250-315 μm) is the most typical technique. A faster technique involves using laser diffractometry on a very small sample (ASTM B822). It is possible to obtain the main particle size parameters of the powder, such as d_{10} , d_{50} or d_{90} (d_x (μm) is the diameter for which the particles

with a dimension less than d_x make up $x\%$ of the total mass, volume or particle number of the analyzed batch; d_{50} is the “median diameter”).

– *Chemical composition*: The ICP-OES method requires dissolving a small amount of powder in an acidic aqueous solution, which is vaporized into the core of an argon plasma induced by a high frequency discharge. The different chemical species are then ionized and thermal excitation generates optical radiations that can be analyzed by a spectrometer. In addition, a measurement of sensitivity to O_2 and H_2O can be useful to assess the influence of the storage and handling atmosphere.

– *Powder morphology*: The spherical or angular nature of the powder particles as well as the presence of satellites can be evaluated by optical microscopy or by SEM, which is associated with image analysis for quantification purpose.

– *Porosity level*: This can be estimated first by means of SEM observations (coupled with image analysis) of particles and polished samples of previously mounted particles (Figure 2.10). The distinction between open pores and closed pores is not easy, because the observation is only in two dimensions and according to a single cross-section through the particle, unless the pores are detected as through pores. Round pores toward the center of the particle can be considered non-through, indicating the presence of gas trapped during atomization. Quantifying the closed porosity volume fraction (non-through pores) of a powder sample is possible by helium pycnometry, which makes it possible to measure with great accuracy the volume of a sample (non-through pores included) of a known mass (DIN 51913). The calculated density is finally compared to that of a fully dense control sample of the same alloy.

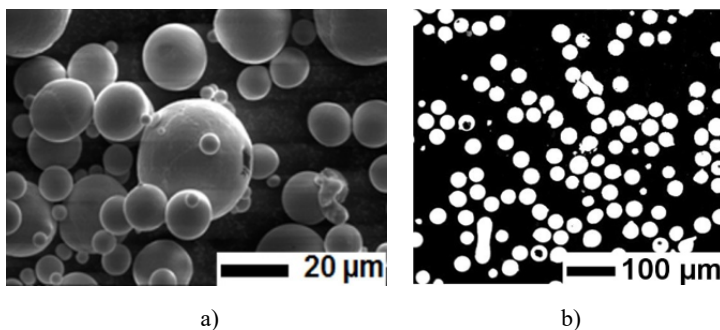


Figure 2.10. Illustration of porous particles: a) SEM micrograph of a particle (in the center) with a through pore; b) SEM micrograph (binarized) of a polished sample showing the pores in the particles (ONERA images)

The through pores can be estimated using SEM observations. The porosity measurements from He pycnometry and SEM are thus complementary. If an in-depth study is necessary, 3D techniques such as X-ray microtomography or the reconstruction of images obtained under a focused ion beam (FIB) can be used, which give access to the nature (through or non-through) and the volume of each of the pores.

– *Apparent (pour) density*: This corresponds to the density of a powder that has undergone a free flow in a container. Various methods can be used, such as the funnel method (ISO 3923-1:2008) or the Scott volumeter method (ISO 3923-2:1981).

2.1.4. Rheological properties of powders

In the field of AM, powders are used in different contexts: for gravity flow through a hopper (static storage or with agitator), pneumatic transport and projection (laser cladding) and formation of a thin powder bed (for L-PBF, E-PBF and MBJ). The mechanical stresses are varied (speed from a few mm/s to a few m/s, stresses from a few Pa to a few bars), as are the sizes of particles (a few μm to a few hundred μm). They are sometimes accompanied by certain problems (discharge of fine particles, blocking of the hopper nozzle, poor or non-uniform compactness of a powder bed, segregation, creation of grooves, etc.). Understanding the rheological properties of powders makes it possible to develop solutions. In this section, we first present the forces to which the particles are subjected and then we compare them. Some key phenomena specific to particulate media are identified.

2.1.4.1. Interparticle forces

Each particle is subjected to different forces which can be classified into: (i) attractive forces (van der Waals, capillary), (ii) repulsive forces (steric repulsion) and (iii) other forces (gravity, electrostatic, solid friction and fluid friction). First of all, every force is detailed below.

– *Van der Waals forces*: Particles behave like dipoles and, as such, are subjected to attractive forces at a distance, the potential of which is $1/r^6$. The force $F(N)$ that attracts two macroscopic objects toward each other is calculated by summing the contributions of each molecule pair. The potential $V(J)$ from which the force derives can be written in the form of a product of the Hamaker constant $A(J)$, which depends on the material, by a function f depending on the morphology of the objects and of the shortest distance h between them:

$$F_{VdW} = \text{grad}(V) \text{ with } V = -A(\text{mat}).f(\text{morpho}, h) \quad [2.2]$$

For two spherical particles of diameter d_{particle} (m), the force intensity is:

$$F_{VdW} = A. \frac{d_{\text{particle}}}{24.h^2} \quad [2.3]$$

The value of the Hamaker constant depends on the solid materials and the surrounding fluid. For objects in air, we have approximately (in $\text{zJ} = 10^{-21}\text{J}$): 70 for silica (SiO_2), 200 for zirconia (ZrO_2), 260 for iron (Fe), 300 for aluminum (Al), 360 for zinc (Zn), 380 for titanium (Ti), 460 for copper (Cu) and 480 for gold (Au). This force is always present, whatever the experimental conditions (Israelachvili 2011). However, in practice, due to the roughness on the surface of the particles, its intensity is lower than theoretically expected.

– *Steric repulsive forces*: This is the Pauli force that is responsible for the non-interpenetration of two solid bodies pressed against each other (notwithstanding their deformability).

– *Capillary forces*: When a liquid is present between two particles, the force of surface tension tends to attract the particles toward each other. In the context of AM, the liquid can be a polymer (MBJ), a bath of molten metal (L-PBF, E-PBF, etc.) or liquid water initially present in the atmosphere. Depending on the amount of liquid, there are five states (Figure 2.11).

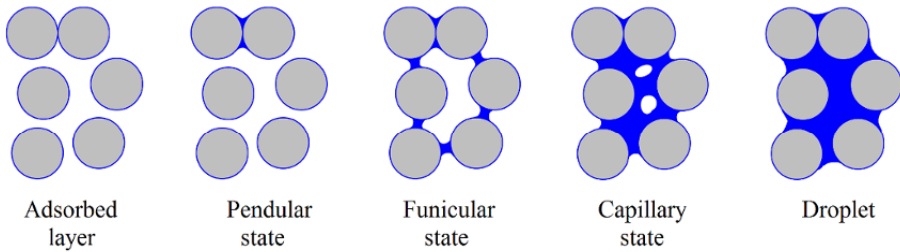


Figure 2.11. States of a particulate medium, depending on the amount of water.
For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

If a powder is exposed to atmospheric air before consolidation (by laser or otherwise), the water vapor is adsorbed to the surface of the particles. The adsorbed layer begins growing with relative humidity ($\text{RH} = p_{\text{H}_2\text{O}}/p_{\text{Sat}}$, with $p_{\text{H}_2\text{O}}$ the partial pressure of water vapor and p_{Sat} the saturated vapor pressure), then transforms into a liquid bridge when RH exceeds a certain threshold. The presence of an adsorbed layer, and *a fortiori* of a capillary bridge, makes the powder cohesive. This major

drawback can disappear if the powder is degased in a dry atmosphere at high temperature or low pressure (sometimes under a vacuum). The time required for evaporation/desorption depends on the temperature and the material; it can be significant (a few hours at 110°C). The cohesion force F_{cap} (N) linked to the existence of a capillary bridge keeping two perfectly spherical particles in contact has the order of magnitude:

$$F_{cap} = \pi \cdot \gamma_{LV} \cdot d_{particle} \cdot K \quad \text{with} \quad K^{-1} = 1 + \tan\left(\frac{\theta}{2}\right) \quad [2.4]$$

where γ_{LV} (J/m²) is the surface tension between liquid and vapor, typically 72 mJ/m² for water and a few hundred mJ/m² for liquid metals. The constant K depends on the wettability angle θ and varies between 0.5 and 1. Quantitatively, pure metals are very hydrophilic: the wettability angle θ is 6° for aluminum (Al), 7° for steel (316) and 8° for copper (Cu) (Trevoy and Johnson 1958). Metal oxides that can be found on their surface are also hydrophilic: 71° for TiO₂ and 60° for ZrO₂ for example. The wettability angle can vary significantly depending on surface roughness and radiation exposure (e.g. UV).

– *Electrostatic forces*: These forces can be either attractive or repulsive. They are due to the electrical charges (positive or negative) which appear on the surface of the particles when they are in contact (instantaneous collision or long friction) with another body (plate, scraper, roller, storage silo, pneumatic transport pipe, etc.). We are talking about tribo-electrification. These forces are not easy to predict, but can have a significant intensity at a long distance (in 1/h²), especially when the surrounding gas does not contain moisture.

– *Solid–solid friction force*: Consider two solid particles in contact under the effect of a normal force F_N . If a weak tangential force F_T is applied, they will stick to each other (adhesion). When the tangential force reaches a threshold value, then a relative movement (sliding) will be observed. Very often, one can apply Coulomb's law which distinguishes two states:

$$\begin{cases} \text{Adherence } (v_{rel} = 0) \Leftrightarrow F_T < \mu_{particle} \cdot F_N \\ \text{Sliding } (v_{rel} > 0) \Leftrightarrow F_T = \mu_{particle} \cdot F_N \end{cases} \quad [2.5]$$

where $\mu_{particle}$ is the friction coefficient, typically between 0.1 and 1. Its value depends on the chemical nature and the roughness of the opposite surfaces, as well as the possible presence of adsorbed water or liquid film. The average normal force

experienced by a particle of diameter $d_{particle}$ (m) constituting a packing of porosity ε subjected to a pressure p (Pa) can be estimated by the formula:

$$F_N \approx \frac{p}{n_c \cdot d_{particle}} \quad \text{with} \quad n_c \approx \frac{8\pi}{3} \cdot \frac{Z(1-\varepsilon)}{d_{particle}^3} \quad [2.6]$$

where n_c is the number of contacts per unit of volume and Z the coordination (the average number of contacts per particle), typically between 5 and 10. For a standard packing subjected to a pressure of 1 bar, we find $F_N \sim 3 \times 10^3 \cdot d_{particle}^2$ (SI units). Since the forces are not isotropic but form a network, a particle can experience a force one or two orders of magnitude higher than the average force.

– *Solid–liquid friction force*: A spherical particle with diameter $d_{particle}$ (m) and density $\rho_{particle}$ ($\text{kg} \cdot \text{m}^{-3}$) in free fall in a fluid is subjected to its weight and to viscous friction with the surrounding fluid of dynamic viscosity μ_{gas} ($\text{Pa} \cdot \text{s}$). In a very short time, it reaches a constant speed, called the terminal velocity U_∞ (m/s), which can be calculated by the formula:

$$U_\infty = \sqrt{\frac{4 \cdot g \cdot d_{particle}}{3 \cdot C_D}} \quad \text{with} \quad \begin{cases} C_D = C_D(Re) \\ Re = \frac{\rho_{gas} \cdot U_\infty \cdot d_{particle}}{\mu_{gas}} \end{cases} \quad [2.7]$$

where g is the gravitational acceleration in $\text{m} \cdot \text{s}^{-2}$ and the friction coefficient C_D depends on the Reynolds number Re . When the latter does not exceed 20, the equation $C_D = 24/Re$ (Stokes regime) can be used. As an example, the fall speed is 2.2 m/s for a 100 μm steel particle and 7.5 mm/s for a 10 μm aluminum particle. These speeds are relatively low. If the gas is driven at a higher speed, then it can easily carry the particles (sometimes where they are not desired). Regarding the comparison of different forces, the net force experienced by a particle is the sum of the individual forces. Figure 2.12 shows the domain of predominance for adhesive forces (Masuda et al. 2006).

Figure 2.12 suggests that in the absence of moisture in the atmosphere, van der Waals forces dominate. However, this must be adjusted. Indeed, in the presence of surface roughness, the van der Waals forces are significantly reduced while the electrostatic forces are not very sensitive to this parameter. It follows that the electrostatic forces can turn out not to be negligible for particles of 10 μm or more, and all the more so when the surface of the particles has been oxidized, even over a thickness of a few angstroms (the oxide layer is electrically insulating).

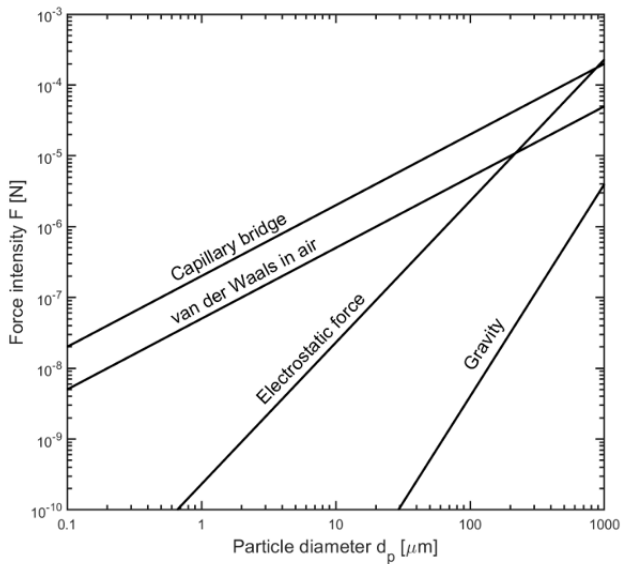


Figure 2.12. Domain of predominant forces acting on a particle

2.1.4.2. Phenomenology of particulate media

The behavior of particulate media is complex: it may look like a flowing fluid or a solid that withstands stresses without significant deformation. The transition between these two states is often difficult to predict and can be the source of undesirable phenomena. Let us cite in particular:

1) segregation: an homogeneous particulate medium composed of two populations of particles (differing in size, density, shape, etc.) can become inhomogeneous when the particles can move relative to each other (shear during spreading in layers, gravitational flow, pneumatic transport by a fluid, agitation by vibration, etc.);

2) agglomeration: when the attractive forces are predominant in relation to the weight (e.g. fine particles), the particles can agglomerate with each other and prevent the formation of a uniform bed during spreading;

3) flow blockage (permanent or intermittent) by forming of arches: this phenomenon manifests itself in silos, pipes and feed hoppers. It can be reduced or eliminated with lower geometric confinement (larger diameter pipe) or by injecting mechanical energy (shocks, vibrations).

2.1.4.3. Rheological characterization tools

A list of characterization tools is given as follows. To obtain an exploitable result (explanatory, predictive, etc.), it is necessary to choose the tool whose measurement principle is closest to the real situation, even if it means inventing an *ad hoc* instrument.

- *Tapped density tester*: This makes it possible to measure the tapped and non-tapped bulk densities (ASTM D7481-09). The Hausner ratio, defined as the ratio of these two densities, is a measure of the compressibility of a powder and, indirectly, its flowability. An index less than 1.25 indicates good or very good flowability and an index greater than 1.40 poor flowability. Although easy to implement, this measurement does not correctly predict the ability of a powder to spread in a uniform layer (Spierings et al. 2016). Example: Quantachrome auto-tap.

- *Flow through an orifice*: The Hall cone (ISO 4490) is a device which measures the time required for a certain amount of powder placed in a 30° half-angle cone to flow (usually 50 g, sometimes 20 cm³ of conditioned powder) through an orifice. This device is generally not suitable for fine powders used in AM because they do not flow through the standard 2.5 mm diameter orifice. Note the possibility of measuring the instantaneous flow by adding a balance, increasing the size of the orifice (Carney's funnel with 5 mm diameter, ASTM B964) or promoting mass flow by reducing the angle (Gustavsson cone with half-angle of 15°, ISO 13517).

- *Angle of repose*: This is the angle between a horizontal plane support and a generator of a conical pile formed by allowing the powder to flow through a funnel (ISO-4490 and ASTM B213). A low angle of repose (<30°) indicates an easily flowing powder while a high angle (>45°) is associated with some cohesion (Geldart et al. 2006), making spreading in layers difficult, if not impossible. Examples: Powder Tester from Hosokawa, GranuHeap from Granutools and so on.

- *Shear cell*: The principle here is to subject a granular stack to a vertical compressive stress, then to apply a horizontal shear stress. This cell makes it possible in particular to measure the internal friction angle, the friction coefficient between the powder and the flat wall of any material (useful variables for assessing a silo and its hopper). It is advisable to choose a low normal stress, like the conditions encountered in AM. Examples: Schulze ring shear tester (ASTM D6773), Jenike shear cell, FT4 from Micromeritics, PFT from Brookfield, MCR from Anton-Paar and so on.

- *Avalanche angle*: A powder is placed in a horizontal axis drum with transparent walls. When rotation occurs, avalanches are observed using a camera. The device characterizes the smoothness of the surface (fractal dimension), the

average avalanche angle and fluctuations about this average. This measurement (and the cohesion index) allows cohesive powders to be well-characterized, although the stress (avalanche caused by gravity) is quite far from spreading by a roller/scrapper. Examples: GranuDrum from Granutools and Revolution Powder Analyzer from Mercury Scientific.

2.1.4.4. *Strategies to improve rheology*

To improve the ability of the powders to flow and spread, we can act on the parameters as follows:

- *Size*: Rheology is improved as size increases, as cohesive forces become proportionally weaker. However, the average size is usually dictated by other considerations (resolution of the finished part).

- *Shape and surface state*: Spherical particles flow more easily than angular or faceted particles. We should therefore choose, in decreasing order of preference, plasma spheroidized powders, gas atomized powders and then water atomized powders. Powders obtained by mechanical fragmentation will *de facto* be more difficult to use. If agglomerates form during consolidation with an energy beam, care should be taken to separate them by sieving before recycling the powders.

- *Size distribution*: For particles that are not very cohesive (therefore not too fine), the mass flow rate at the outlet of a silo increases slightly (up to approximately 10%) when small particles are added (ball bearing effect) (Benyamine et al. 2014). Furthermore, during laser sintering, particles close to the beam can agglomerate by sintering and form large clumps; recycling must then be accompanied by sieving, otherwise the quality of subsequent layers will deteriorate.

- *Surface chemistry*: Two techniques can be used to modify the surface finish, coating (by PVD, CVD, etc.) or heat treatment. At moderate temperatures (110–200°C), a fraction of the water adsorbed to the surface will desorb and thus reduce capillary forces. At higher temperatures (500°C), the composition of the surface oxide layer can change and induce an evolution in the acid–base behavior, which can significantly improve the flowability of a powder (e.g. Ti6Al7Nb, $d_{50} = 34 \mu\text{m}$).

- *Vibrations*: The mechanical energy input, by vibrating the container filling with powder, locally and temporarily “breaks up” the agglomerates due to cohesive forces as well as the arches and vaults attributable to the Janssen effect. This significantly improves the flow speed in a hopper and increases its evenness. In general, high frequencies are required (from 1–50 kHz).

2.1.4.5. Modeling of particulate flows by distinct/discrete element method

To numerically simulate the behavior of a set of particles, the most widely used method is the distinct/discrete element method (DEM). Each real particle is modeled by a slightly deformable sphere or an assembly of such spheres (Figure 2.13). Knowing the total force exerted on each sphere, the fundamental principle of dynamics makes it possible to calculate its acceleration and then, by integration, its speed and its position. In this way, we have access to microscopic quantities that are difficult to measure experimentally: the trajectory of each particle and the forces to which it is subjected.

The total force to which a particle is subjected is the sum of two contributions: (i) distance forces, which include weight, electrostatic forces and van der Waals forces; and (ii) contact forces, which include forces of solid and viscous friction, steric repulsion and capillary forces. All these forces have been described in section 2.1.4.1.

Although powerful and very useful, this method often requires very long computation times, even by resorting to parallelization, because its explicit nature imposes very small time steps (10–100 μ s). Moreover, to calibrate the numerous numerical parameters, it is necessary to experimentally characterize the behavior of the particles taken individually or as a whole.

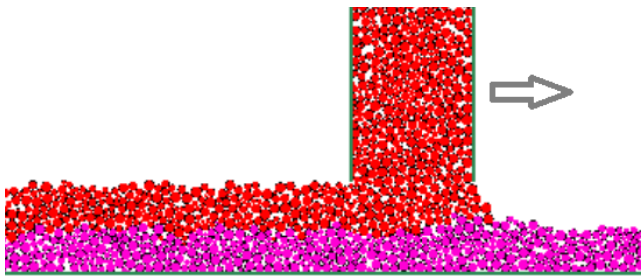


Figure 2.13. Layering by gravity feed. For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

Among the preferred applications are powder transport (gravity or pneumatic, with or without vibrations) and layering using a scraper. These numerical simulations provide a better understanding of the conditions under which there may be blockages in the flow, irregularities in a bed, heterogeneity of local compactness, a stress field, etc. Simulations with the DEM method which incorporate heat transfers from the literature are described below.

2.1.5. Influence of powders on processes and final properties

It is very important in AM to respect very precise specifications for powders (particle size, morphology, chemistry) to guarantee repeatability of mechanical properties in finished parts. This inevitably increases production costs and tends to restrict these powders to high added value markets (medicine, aeronautics). As such, it is recommended to use atomization processes favoring the use of a protective atmosphere and the spheroidization of the powder particles without trapping any internal gas, which also makes it possible to control the size of the particles when modifying the gas flow/metal flow ratio.

However, work is underway to study the possibility of using less restrictive processes for AM, such as, for example, water atomization for certain non-reactive powders, with a drying step and often a deoxidation treatment. Apart from its reduced cost, water atomization has the advantage of high variability in water jet pressure, making it possible to vary the size and morphology of the powders. Despite the disadvantage of obtaining irregular shaped particles, a very good density can be obtained by modifying the parameters of the AM process, thus leading to satisfactory mechanical properties. These promising results seem to pave the way for a “democratization” of conventional powders for AM; however, the requirements and costs required for the raw materials in order to guarantee the desired properties for the finished parts are still uncertain.

2.1.5.1. Influence of particle size

With regard to powder bed processes, it is necessary for the size distribution of the powder particles to guarantee the best possible compactness in order to ensure good layering. However, the impact of deviating from an optimized granulometric class on build quality has not yet been thoroughly investigated. Naturally, it is clear that using a powder with a relatively large particle size will require thicker layering than with a fine powder. In this case, since the particle intervals are larger, one can expect a higher porosity rate for the laser-fused material (Tan et al. 2017). It is possible to deduce from this that a low layer thickness and a fine powder are preferred in order to avoid porosities in the finished parts, but to the detriment of the manufacturing efficiency. In addition, this extrapolation is limited, because the use of a too fine powder whose granular behavior is no longer controlled by the gravity forces leads to poor flowability and, consequently, to an increased risk of surface defects when spreading the powder (Qiu et al. 2015). In summary, the choice of particle size is made on the basis of a compactness–flowability compromise.

2.1.5.2. Influence of particle morphology

It was already mentioned in section 2.1.4.4 that an angular or faceted contour for powder particles does not contribute to good flowability. Likewise, the presence of elongated particles and the presence of agglomerates will also slow down or block the powder flow. Once again, a Hall's funnel should not be the only apparatus used to qualify the flow of any particular batch of powder. A fairly convincing illustration was given by Clayton et al., who compared three batches of stainless-steel powder (A, B, C) from the same supplier, the three batches having the same powder size distribution and the same results in Hall flowability (Clayton et al. 2015). During laser deposition tests on these three separate batches, it turned out that the C powder exhibited significant mechanical friction, which induced a very significant reduction in the mechanical properties of the lasered parts. Complementary measurements based on other experimental means (Freeman FT4) have made it possible to establish convincingly a strong correlation between the specific energy of castability and the properties of the samples lasered for these three batches of powder. It is thus desirable to show in the powder batch identification sheets the various measures, which may relate to their flowability. In addition, a better prediction can undoubtedly be made through a precise quantification of the powders' morphology (sphericity, circularity) and the establishment of a correlation between this morphology and the main criteria for measuring the flowability.

2.1.5.3. Tolerances of chemical elements

The influence of the powder's initial chemistry on the final properties of the parts is linked to the process and in particular to intense thermal phenomena during manufacture. Thus, evaporation and volatilization phenomena of certain elements with a low vaporization point can contribute to significant differences in the chemical composition between the powder and the lasered part. For example, 6061 aluminum (Al-Mg-Si) powders can experience significant loss of Mg and Si elements during manufacture. This deviation in chemical composition negatively affects the alloy, making it more susceptible to cracking. For this alloy, it is therefore recommended to adjust the chemical composition of the initial powder, so as to be able to anticipate the subsequent loss of these elements under the laser beam.

The case of the AlSi10Mg alloy is interesting to study here in that the element contents can vary significantly from one powder supplier to another. For example, a high Si content is favorable for laser absorption as well as surface tension and therefore fluidity. The downside is that a composition that is too close to the eutectic tends to promote the formation of platelets, which are cyclic fatigue nucleation sites. This has led some researchers to instead modify the L-PBF process conditions for

the hypo-eutectic alloy despite a less spherical morphology for this powder (Aboulkhair et al. 2015). Thus, a strategy with pre-sintering at 50 W then melting at 100 W allowed them to obtain good density and good properties.

Variability in powder chemistry has also been identified to have a strong impact on the final product in the case of steels (Averyanova et al. 2012). In this case, a comparison of two batches of stainless-steel powder (17-4 PH) with precipitation of Cu-rich nano-precipitates was made. The differences observed in the contents of C and Ni, which are elements stabilizing the γ phase, induced microstructural differences, with powder 1 entirely martensitic and powder 2 composed of 86% γ austenite phase. It is useful to specify that the particle size characteristics of the powders were also different: powder 1: $D_{90} < 16 \mu\text{m}$ and powder 2: $D_{90} < 50 \mu\text{m}$. Post-fabrication results revealed that the type of powder had a strong effect on phase transformations during thermal cycling. Powder 2, which was enriched in C and Ni elements, indeed led to a significant amount of residual austenite after L-PBF (94% compared to 62% for powder 1).

The atmosphere and the gas pressure used inside the machines must also be taken into account and depend on the nature of the chemical species present in the powders. An air atmosphere is therefore to be avoided for materials liable to form undesirable nitrides. On the other hand, vacuum processes such as the E-PBF process have the disadvantage of promoting aluminum loss for alloys that contain it, in particular titanium alloys (Ti6Al4V, TiAl). In addition, these evaporation phenomena are extremely local in that they occur at the top of the molten layers. As a result, there is an aluminum content gradient in the layers. These local variations in aluminum concentration modify the solidification path, leading to undesirable variations in the size and morphology of the microstructure along the Z orientation.

2.1.5.4. Influence of powder recycling

Due to the relatively high cost of the raw material, it is rare nowadays for parts to be made only with new powder. However, a powder not used in the manufacturing chambers is not necessarily free of traces of contamination and cannot be considered identical to a new powder.

This is a real dilemma for the aviation industry in particular. On the one hand, it is not very profitable to dispose of unused powder because of its cost, despite the fact that reuse leads to potential pollution risks; on the other hand, the quality and performance requirements may mean not recycling the powder due to the higher number of defects in a powder that has already been used.

In general, the composition of the powder, the particle size distribution, the bulk density, the tapped density, the apparent density, the flowability and the morphology of the particles are all characteristics to be studied when the powders are reused. More particularly, it should be noted that the specific surface area of powders is approximately 10,000 times greater than the surface area of an equivalent solid volume. Consequently, the issue related to recycling powders concerns the effects of surface contamination. Hryha et al. (2010) used different advanced techniques for a variety of metallic materials, such as an FIB to assess the thickness of the oxide layer, X-ray photoelectron spectroscopy (XPS) to obtain information on the chemical composition of the oxide layer and its thickness, and Auger spectrometry to determine the chemical composition of the oxide with high resolution (10 nm). Thus, for steels, increasing the Cr, Mn and Si contents tends to increase the size of the oxide particles on the surface. On the other hand, fresh IN718 powders tend to be coated with a thin layer of stable Ni oxide, while powders recycled five times are coated with Cr- and Al-rich oxide particles (Hryha 2017). The recycling of Ti6Al4V powder in EB-PBF is also well-documented (Tang et al. 2015). It turns out that the morphology of the powders changes significantly during recycling, with a deformation of the particles due to the preheating step at 730°C and also due to the impacts in the powder recovery system (PRS). In addition, the particle size distribution tends to narrow with the gradual disappearance of satellites and agglomerates by sieving. An important effect of recycling is the increase in the O₂ content from 800 to 1,900 µg/g after 21 recycles. Concomitantly, an increase in yield stress (YS) (and ultimate tensile strength (UTS)) from 834 MPa (920 MPa) to 960 MPa (1,039 MPa) is observed in parts with powder recycled 21 times, the ductility remaining constant (15–16%).

2.1.6. Standardization

The general proliferation of AM technologies is still hampered by a lack of standardization. Standards are essential to ensure the necessary confidence in the level of accuracy and reliability of measurements and good practices. The industry can already count on a number of standardization initiatives from the following organizations:

- ASTM International, Committee F42 (16 countries, 200 organizations, established in 2009);
- ISO Technical Committee 261 (17 countries, established in 2011);
- AFNOR/UNM 920 Committee (established in June 2010).

In the particular case of metallic powders, we can point out the recent efforts to standardize the characterization methods, generally derived from industrial

practices, within the joint framework of ASTM International and the International Organization for Standardization (ISO), with the release in 2014 of the standard F3049-14: Standard Guide for Characterizing Properties of Metal Powders Used for Additive Manufacturing Processes. This standard identifies 37 ASTM, ISO and MPIF methods useful for characterizing metal powders. They include methods of sampling and of measurement of powder size, morphology, chemical composition, density and flowability.

Mention should also be made of the French standard NF E 67-010 (12/19/2014) Additive Manufacturing – Powders – Technical Specifications, which deals with the technical conditions for delivery of powders of metallic, plastic and ceramic materials and specifies the information to be provided by the powder supplier. This standard specifies in particular the characteristics that are required and those that remain optional.

2.1.7. Summary

Metal powders that are used in AM should probably not be seen as a simple raw material but rather already as a semi-finished product that is shaped for optimal use in AM machines. However, the methods or tools currently used to measure powders in AM are not sensitive enough or appropriate to obtain the powder characteristics that are important during a given AM process. As good practice, it is recommended to combine several types of measurements to truly understand, for example, the flow behavior of a powder.

We have seen in these pages that excessive variations in the characteristics of the powders lead to non-uniform layering, an increase in defects, density differences of the bulk and degraded surface finishes, and thus to a reduction in mechanical properties. It is also important to assess the degree of harmfulness of each type of defect present in the powders and threads with respect to potential applications. It should also be remembered that the characteristics related to the powder chemistry (loss of elements, interstitial contaminants) depend on the type of material and the type of AM technology.

In the future, it will be important to ensure increased quality control of powders during transport and storage, for example by incorporating a number of sensors to measure humidity, temperature, pressure, vibration and oxygen levels. Finally, with regard to the characterization methods, the powders are also suitable for the use of sophisticated imaging coupled with machine learning.

2.2. Metal wires

2.2.1. Introduction

In recent years, technologies based on metal wires derived from welding have been used for the creation of 3D materials. Far from competing with processes that use powders as the raw material, the specific characteristics that are offered allow them to be considered as complementary. We can even consider, for certain part geometries, using a hybrid powder/wire process. The main characteristic of using metal wires as a raw material is the need for a development step to make the wire into lengths suitable for layer-by-layer deposition. A direct consequence can be observed in the handling of this raw material by the operators, in that the metal wires are safer than metal powders. Another major advantage of using wire relates to the speed or rate of deposition, which is greater than that of powder deposition processes. Depending on the chosen deposition process, the rates vary from 500 g/h (wire arc additive manufacturing (WAAM) tungsten inert gas (TIG) deposition) to several kilograms per hour (WAAM metal active gas (MAG) deposition or electron beam additive manufacturing (EBAM) deposition). The wires used have a diameter that commonly varies between 0.8 and 1.6 mm. However, the surface finishes are generally rougher, which may require a finishing step. On the other hand, significant overheating of the part occurs in multipasses, which causes a significant level of residual stresses and, consequently, problematic deformations vis-à-vis the dimensional tolerances. To overcome this geometric problem and also help obtain the desired microstructures in the parts, researchers are focusing on modelling, but also at the experimental level on optimizing the trajectories.

In this section, we provide some illustrations of the use of metal wires in AM with some major French stakeholders. However, beforehand, it is appropriate to provide a quick summary of the three best-known processes using metal wire as the raw material. These processes actually differ in the energy source which can be an electric arc, a laser or an electron beam. (Ding et al. 2015).

WAAM involves the use of arc welding equipment and allows for high deposition speeds leading to high deposition rates (4 kg/h). With metal inert gas (MIG)/MAG technology, the wire is used directly as a consumable electrode to generate an electric arc, while with TIG and plasma technologies, the electric arc is formed through a refractory electrode. The wire feed is coaxial with the nozzle in the case of the MIG/MAG welding technique, which facilitates the management of

the trajectories and allows high deposition speeds. Nevertheless, TIG and plasma techniques are widely used with titanium in the aeronautical sector thanks to the high stability of the electric arc. Cranfield University was a pioneer in making MIG/MAG steel and aluminum parts, but also in developing large-scale titanium alloy structures using TIG, notably for Thales Alenia Space. The latter also adopted Norsk Titanium plasma technology, with the production of the first FAA-approved titanium part for the Boeing Dreamliner (Norsk Titanium 2017). For its part, STELIA Aerospace has produced aluminum fuselage panels with stiffeners using the WAAM process (DEFACTO project). Other players have demonstrated the interest of the WAAM process, especially for marine applications. Thus, the joint laboratory formed by Naval Group and Centrale Nantes produced, using the WAAM process, the first large military propeller blade made of copper alloy (over 300 kg) with an innovative design. This hollow propeller concept, resulting from the H2020 RAMSSES project, makes wire deposition dependable compared to casting for large parts, with a reduction in the time and cost of manufacturing parts.

Wire laser additive manufacturing (WLAM) can be used for different types of metal materials, but the operating conditions are specific to each material, depending in particular on its absorptivity vis-à-vis the laser beam. The process involves forming a weld pool on the surface of a substrate, which may be an existing part, so as to fuse the filler wire and thereby build up a deposit. The operation can be repeated on several layers to, for example, renew a used part or add a function. The most influential parameters are the laser power, its displacement speed and the wire flow. The deposit will thus be greater if the displacement speed is reduced and if the wire flow rate is increased; but the laser power must then be adjusted to increase the temperature of the weld pool in order to avoid an excessively viscous melt or even incomplete melting. Other parameters are also important such as the wire-substrate angle and the position of the wire in front of the laser beam to promote even wire deposition. The laser beam must itself be slightly inclined with respect to the normal of the surface of the substrate to avoid any risk of reflection on the optics. As an application example, GKN Aerospace used the laser wire deposition technique for the manufacture of the Vulcan 2 nozzles for Ariane 5.

The electron beam-based process, called electron beam free form fabrication (EBF3) or electron beam additive manufacturing EBAM, is carried out in a vacuum chamber. The main parameters are identical to the two previous processes: the power generated by an electron beam, the deposition speed and the wire feeding speed. These three parameters affect the energy density and thus the quantity of melted material and the geometry of the deposit. If a high deposition rate is favored for obtaining large parts, a fairly large deposition width is favored, which

is then not compatible with obtaining fine geometric details. Sciaky is the first company to have developed a number of machines and has demonstrated the feasibility of manufacturing large titanium parts for Lockheed Martin with good compliance and satisfactory mechanical properties.

2.2.2. Wire production

An example of wire production is given in the following with the manufacturing of alloy 625 (UNS N06625) developed by Aperam (Figure 2.14), which has characteristics adapted to market requirements (welding and coating), that is to say, integrated high-quality production ensuring responsiveness to requests. The different steps for obtaining the wire rod are as follows:

- development of new materials in order to obtain an alloy containing 0.5%_{mass} less iron;
- arc furnace fusion, with new refractories to avoid pollution;
- vacuum oxygen decarburization (VOD) refining process in order to carry out the three main operations:
 - decarburization by blowing oxygen and vacuum pumping (several mbar),
 - lime-based deoxidation and desulphurization in slag,
 - adapted concentrations of the reducing elements Nb, Al and Ti;
- continuous rotary casting (CRC) in billets 150 mm in diameter;
- reheating of the billets in a gas oven;
- roughing and hot rolling to 5.5 mm in diameter (wire rod) on a wire train, followed by hyperhardening;
- heat treatment to prepare the surface for the pickling operation and develop the mechanical properties required by wire manufacturers;
- stripping and quality control (dimensions and tracking for the absence of defects such as scales, fibers, folds, etc.).

The typical chemical composition (in mass%) obtained by wire rod manufacturing is given in Table 2.2. Note the very low content of residual S, P and Ca, ensuring excellent weldability for the future wire, an essential use value in welding but also in AM.

Ni	Cr	Mo	Fe	Nb	C	Si	Mn
Equal	22	8.8	0.35	3.6	0.01	0.4	0.03
Ti	Al	Mg	Ca	S	P	N	
0.06	0.14	0.005	0.0006	<0.0005	<0.003	0.015	

Table 2.2. Chemical composition (%mass) of a 625 alloy developed by Aperam Alloys

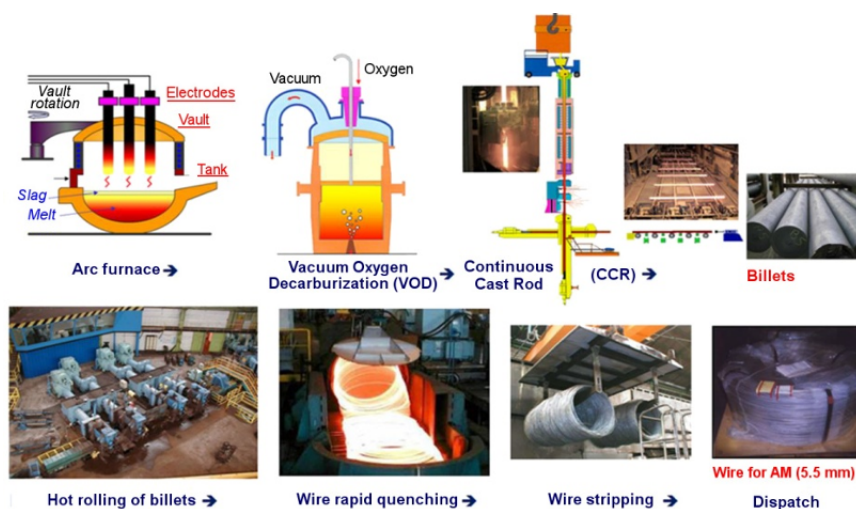


Figure 2.14. Production of a wire rod with alloy 625, diameter 5.5 mm (Aperam Alloys)

A wire rod (diameter 5.5 mm) with alloy 625 is then available:

- either to supply the welding companies that will carry out the wire manufacturing, making it possible to supply the welding and coating market, but also that of AM, with MIG wires (welding wire conditioned for automatic welding);
- or to supply plasma atomization frames, after a light manufacture, in order to produce high quality micrometric powders.

Alloys	Ni	Co	Cr	Mo	Nb	W	Cu	Other elements
Imphy N99	>99	–	–	–	–	–	<0.1	Fe < 0.2
Imphy CuNi30	30/32	–	–	–	–	–	65/69	Ti 0.2/0.5 – Fe 0.4/0.7
Imphy NiCu28Ti	63/68	–	–	–	–	–	28/29	Ti 1.5/3.0 – Fe < 0.25
Imphy 25-9-4	9/10	–	24/27	3.5/4.5	–	–	<0.3	N 0.2/0.3 – Fe Eq.
Imphy NCW	Eq.	–	21/23	9.5/10.5	–	2.5/3.5	<0.1	Ti < 0.25
Imphy 22	Eq.	–	21/22.5	12.5/14	–	2.5/3.5	<0.2	Fe 4/5
Imphy 276	Eq.	–	15/16.5	15/17	–	3.2/4	<0.2	Fe 4.0/7.0
Imphy 686-1	Eq.	–	21/21.5	15/17	–	3.0/4	<0.2	Al 0.15/0.35
Imphy 625	>62	–	21.9/22.5	8.5/9.5	3.5/3.8	–	<0.3	Ti 0.1/0.25 – Fe < 0.5
Imphy 82	Eq.	–	19.5/20.5	–	2/3	–	<0.1	Ti 0.25/0.45 – Fe 0.75/1.5 – Mn 3
Imphy 718	>51	–	17/18	2.8/3.3	5/5.4	–	<0.1	Al 0.4/0.6 – Ti 0.8/1.15 – Fe 19/21
Imphy X	Eq.	1.5	22	9	–	–	<0.3	Fe 18 – W 0.6
Imphy 617	Eq.	11–12	22/23	8.5/9.5	–	–	<0.1	Al 1/1.5 – Fe < 1 – Ti 0.2/0.5

Table 2.3. Chemical compositions (% by mass) for types of alloys produced by Aperam Alloys in the form of wire rods

The production range implemented with other alloys, in the form of wire rod and dedicated to the welding and coating market, is just as demanding. The grades presented in Table 2.3, as an example, are available for the AM market.

The customer could thus benefit from a range of raw material products (powders plus wires) from the same casting in order to feed hybrid machines (e.g. laser powder/laser wire). Voestalpine is also involved in WAAM with the marketing of a wide range of filler wires dedicated to AM. The Böhler 3Dprint range marketed by Voestalpine Böhler Welding, a recognized manufacturer of filler metals for welding, covers many grades of non-alloy and low-alloy steels, stainless steels, nickel alloys, titanium and aluminum. These filler wires have been developed to meet the specific needs of arc-wire AM.

2.2.3. Use of filler wires in AM

2.2.3.1. Details and considerations for filler wires

Analogies between conventional welding and cladding processes and arc-wire AM are obvious and well known. However, arc-wire AM requires even greater attention to a number of parameters, in particular filler wires.

The layer-by-layer deposition method leads to the formation of columnar structures of different thicknesses depending on the geometry of the part to be manufactured.

In WAAM, for thin structures, the heat is propagated unidirectionally, while in conventional welding it is bi- or even tri-directional. This aspect is essential because it has a great influence on the cooling rates and on the residual stresses, which have consequences on the properties of the deposited metal. Added to this is the need to deposit the metal layer by layer without completely remelting the bottom layer with each new pass, and without risking the collapse of the build-up structure. These limitations have therefore led to the use of lower energy processes than the standard MIG/MAG process, such as the cold metal transfer (CMT) process developed by Fronius.

The viscosity of the melt pool also plays a big role, as this pool needs to balance over previous layers, without running down either side of the bead. It is also difficult to ensure optimal gas protection because of the absence of edges to confine the protective atmosphere. For this reason, the WAAM process is more sensitive to the formation of porosities, oxides and silicates. High arc stability is beneficial, a less

erratic arc allowing more effective protection of the melt pool. The negative impact on the integrity of the material can also be deleterious for the stability of the process and for its efficiency during the deposition of successive layers.

Another difference with conventional welding is the more frequent use of heat treatments in AM to reduce residual stresses within the part, in particular for structural fatigue. The filler wires developed for this process must therefore allow heat treatment to be carried out without any detrimental effect on mechanical properties of the parts.

Furthermore, the constant search for productivity leads to the greatest possible increase in deposition rates and therefore in heat input. Also, for productivity reasons, it is also crucial to minimize the number of stops and to increase the life of the constituent parts of the reel such as the sheaths and contact tubes, which requires a perfectly homogeneous wire surface in order to limit the abrasion phenomena and the formation of residues as much as possible in the feeding system. Finally, for obvious reasons relating to the feeding efficiency, the wires must have sufficient stiffness. This may raise a particular problem in the case of aluminum wires, a solution being to use notched pinch rollers to ensure movement.

2.2.3.2. Optimization of compositions

In addition to the immense engineering work carried out on the optimization and development of increasingly efficient metal AM processes, the metallurgical industry, which manufactures welding wires, is involved in the development of new chemical compositions in order to meet the functional and technological requirements of AM. The industrial strategy adopted by Aperam, which seems to be preferred among the producers of welding wire, is reasonable:

- initially, they supply the AM market with industrial wire defined by the standards and codes in use (e.g. AWS). This makes it possible to build parts with the morphological characteristics specified by the design consultants: dimensional accuracies, surface quality, absence of internal defects, optimized topology, manufacturing productivity, etc. These specifications are the advantage of AM workshops. The industrial wire available on the market is used as a “proof of concept”;

- then, they take into account the specificities of AM characterized by various local mechanisms such as melting, solidification, heat reallocation, sintering, mixing, etc. In other words, they develop new alloy compositions adapted to the use

value of the parts (properties) but also to the various AM processes (NF EN ISO/ASTM 52900, NF EN ISO 17296): material projection, binder projection, powder bed fusion, material extrusion, material deposition under concentrated heat flow and stratification of layers.

For example, to produce filler wires that can withstand various manufacturing constraints, Voestalpine Böhler Welding uses very precise chemical compositions, with fewer impurities to limit the formation of oxides and, for some alloys, less silicon than for conventional welding consumables in order to limit the fluidity of the melt pool. These wires make up the 3Dprint range (Table 2.4).

2.2.4. Influence of wires on processes and final properties

2.2.4.1. WAAM process using different filler wires

The filler wires of the 3Dprint range offer good arc stability and good feeding through an improved surface texture, obtained by applying particular drawing speeds and by depositing specific lubricants on the surface of the wires, different from those applied to the welding wires. For copper wires, optimizing the process parameters makes it possible to obtain a more homogeneous copper plating, with better adhesion.

Finally, concerning the winding of the wires at the time of processing, whether on reels or in barrels, tighter tolerances should be observed in order to optimize the remanence values and left winds, and therefore the straightness of the wire at the torch outlet and the stability of the point of impact.

Comparative tests have shown a much smaller dispersion of parameters with a wire from the 3Dprint range, manufactured according to these technical requirements, compared to a conventional welding wire, which confirms its greater performance stability.

In addition, the possibility of estimating the properties of “printed” parts in advance is an important issue in the arc-wire AM approach. Due to the absence of standards dedicated to AM – at least with regard to the characterization and classification of materials – which would reflect current knowledge, we can consider that the fundamental properties of filler wires are accessible according to specific standards for conventional welding.

As mentioned previously, the type of structure and the cooling rate can vary greatly between conventional welding and AM. One might therefore wonder, for example, whether these parameters affect the tensile strength of the deposited metal.

Reference	Chemical Composition (%)									
3Dprint AM 35	C	Si	Mn							
	0.10	0.30	1.05							
3Dprint AM 46	C	Si	Mn							
	0.10	1.00	1.70							
3Dprint AM 50	C	Si	Mn	Ni						
	0.10	0.65	1.40	1.35						
3Dprint AM 62	C	Si	Mn	Ni	Mo					
	0.10	0.65	1.55	1.10	0.40					
3Dprint AM 70	C	Si	Mn	Cr	Ni	Mo				
	0.08	0.60	1.70	0.20	1.50	0.50				
3Dprint AM 80 HD	C	Si	Mn	Cr	Ni	Mo				
	0.09	0.40	1.70	0.35	2.00	0.60				
3Dprint AM P22	C	Si	Mn	Cr	Mo					
	0.08	0.5	1.0	2.50	1.00					
3Dprint AM 304L	C	Si	Mn	Cr	Ni	N				
	0.02	0.50	1.60	20.00	10.00	<0.10				
3Dprint AM 316L	C	Si	Mn	Cr	Ni	Mo	N			
	0.015	0.45	1.60	18.50	12.00	2.60	0.04			
3Dprint AM 2205	C	Si	Mn	Cr	Ni	Mo	N			
	0.025	0.50	1.30	22.00	5.00	3.00	0.14			
3Dprint AM 2209	C	Si	Mn	Cr	Ni	Mo	N			
	0.025	0.50	1.60	22.00	9.00	3.00	0.14			
3Dprint AM 15-5PH	C	Si	Mn	Cr	Ni	Mo	Cu	Nb		
	0.04	0.50	0.55	14.6	4.8	<0.30	3.40	0.28		
3Dprint AM 17-4PH	C	Si	Mn	Cr	Ni	Cu	Nb			
	0.02	0.35	0.45	16.3	4.60	3.30	0.25			
3Dprint AM 625	C	Si	Mn	Cr	Mo	Ni	Nb	Fe		
	<0.03	<0.25	<0.25	22.0	9.0	Balance	2.50	<2.0		
3Dprint AM 718	C	Si	Mn	Cr	Mo	Ni	Nb	Al	Ti	Fe
	0.05	<0.10	<0.10	17.60	3.00	Balance	5.20	0.45	0.95	19.5
3Dprint AM 430	C	Si	Mn	Cr						
	0.07	0.80	0.70	17.50						
3Dprint AM 410NiMo	C	Si	Mn	Cr	Ni	Mo				
	0.01	0.65	0.70	12.20	4.80	0.5				
3Dprint AM Al 2219	Mg	Mn	Cu	Zr	Ti	Al				
	<0.02	0.35	43165	0.18	0.14	Balance				
3Dprint AM Cu 6328	Mn	Ni	Fe	Al	Cu					
	1.0	4.50	3.50	9.0	Balance					
3Dprint AM Ti-5	C	Fe	O	H	Al	V	Ti			
	0.05	<0.15	0.18	<0.01	6.0	4.0	Balance			
3Dprint AM Tool 40	C	Si	Mn	Cr	Mo	Fe				
	0.1	0.4	0.6	6.5	3.3	Balance				
3Dprint AM Tool 45	C	Si	Mn	Cr	Mo	Ti	Fe			
	0.25	0.5	0.7	5.0	4.0	0.6	Balance			
3Dprint AM Tool 55	C	Si	Mn	Cr	Mo	Ti	Fe			
	0.35	0.3	1.2	7.0	2.0	0.3	Balance			

Table 2.4. Typical chemical composition (%_{mass}) of filler wires for metal additive manufacturing in the 3Dprint range developed by Voestalpine Böhler Welding (note: the alloys for which the scale is not specified are iron-based alloys)

Several studies have been carried out on different alloys with 3Dprint wires manufactured by Voestalpine Böhler Welding in order to assess the differences between the mechanical characteristics of simple parts produced by AM, mainly of the wall type, compared to the properties of the metal deposited without dilution (apart from the mixture with the substrate) in conventional welding.

Different alloys were tested during these studies: non-alloy steel, high elastic limit steel, austenitic stainless steel, duplex, aluminum bronze, etc. In general, for fine structures produced by AM, there is a difference in tensile strength. This is linked to the unidirectional nature of the heat propagation within this type of structure. When the thickness of the parts increases, the heat spreads in several directions and the tensile strength becomes close to that obtained in conventional welding.

However, certain studies have shown that it is possible to achieve mechanical properties similar to those obtained by welding on metal deposited without dilution (that is to say without the contribution of a possible substrate) because of the adapted deposit process. It is therefore certain that, if selecting the grade of the filler wire and the constraints associated with the arc-wire AM process have a significant influence on the characteristics of the resultant parts, an adjusted deposition procedure counteracts negative effects and achieves results comparable to those obtained in welding, as experiments show.

2.2.4.2. Specific features for the different families of alloys

As an example, let us mention a few questions at the heart of the business of wire suppliers for welding and metal AM in connection with the different families of alloys.

2.2.4.2.1. AM of superalloy parts (Ni-Fe-Cr-Mo-Nb-Ti alloys)

According to the Scheil–Gulliver model (Schaffnit 2015), the partition coefficients k_s illustrate the propensity of certain chemical elements to segregate within interdendritic spaces. Typical values to consider in the alloys of the 625 family (UNS N06625) are $k_s = 0.3/0.5/0.7/0.8$ for the elements C/Nb/Ti/Mo, respectively. Segregation may result in the precipitation of undesirable intermetallic phases (TCP, Laves, σ), making the superalloy sensitive to shocks. In such a situation, if the chemical homogenization and the resolution of the phases at approximately 1,150°C is difficult to achieve, for technical reasons (large part, creep, etc.), it is necessary to increase the solubility of the elements constituting the phases by modifying the concentration of certain major elements. The examples presented in Figure 2.15 illustrate the problem involved.

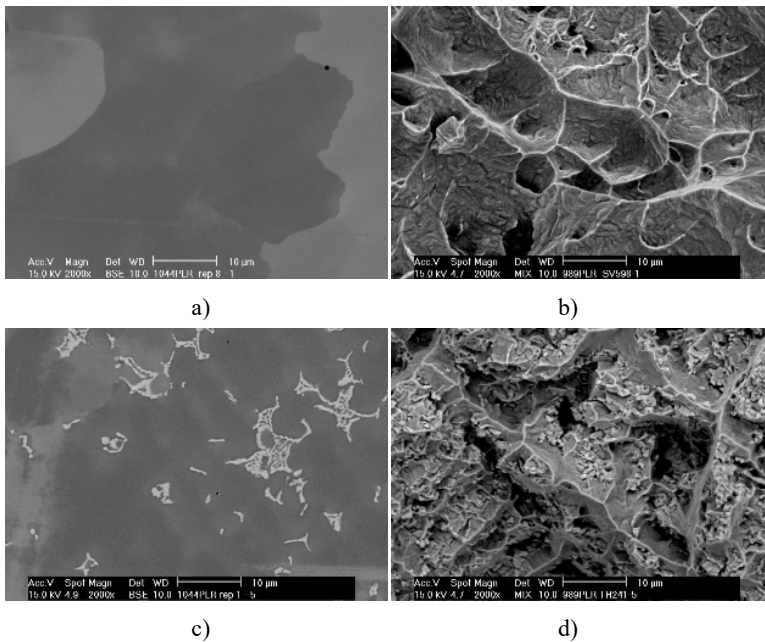


Figure 2.15. Solidification structures (a, c) and failure facies (b, d) observed with SEM on impact test specimens from TIG welds carried out on two model Ni-Cr-Mo type 625 alloys (Aperam results)

COMMENT ON FIGURE 2.15.– (a, b) Solidification structure devoid of interdendritic phase and ductile rupture ($KcV \sim 190 \text{ J/cm}^2$). (c, d) Solidification structure rich in interdendritic phases and brittle fracture ($KcV \sim 20 \text{ J/cm}^2$).

2.2.4.2.2. AM of maraging steel parts (Fe-Ni-Mn-Cr-Mo-Ti-Al alloys)

High yield strength alloys are martensitic hardening grades used for their high mechanical properties combined with excellent ductility. These result from a complex microstructure developed during hot processing (Galindo-Nava et al. 2016): martensite slats (α') with a nanometric precipitation of hardening intermetallic phases (NiAl , Ni_3Ti , Ni_2AlTi , etc.) accompanied by reversion austenite (to regulate ductility). The flow stress is a function of the size of the martensite slats (Hall–Petch mechanism) conditioned by the size of the austenitic grains obtained during the austenitization treatments. However, the solidification structures developed in AM are crude. Can we talk about austenitic grain size, and what would the mechanical properties be? One solution to the degradation of mechanical properties will likely be to increase the precipitated volume fractions. The same

reasoning holds true for superalloys. The question of their possible deterioration during hot cracking, which is well known in alloys with high Ti and Al content, should be considered (DuPont et al. 2009).

2.2.4.2.3. AM of duplex stainless-steel parts (Fe-Cr-Ni-Mo-N alloys)

The mechanical properties and resistance to aqueous corrosion of duplex stainless steels depend on the adjustment of the duplex microstructure (Charles 2010) and the ratio $[\alpha]/[\gamma]$. However, current mechanical parts are mainly derived from conventional metallurgy based on thermomechanical transformation. As the duplex structure of an “as-solidified” metallurgical state differs from that of a wrought state, it is also necessary to develop chemical compositions of wires with modified levels of Ni, Cr and Mo to readjust the ratio $[\alpha]/[\gamma]$.

2.2.4.2.4. AM of INVAR or permalloy parts (Fe-Ni-Co-Cr-Mo alloys)

The thermal expansion of INVAR or the magnetic permeability of permalloys pass through an extremum depending on the nickel content, up to 36% Ni and 78% Ni, respectively. As a result, a segregated nickel structure will develop thermal expansions or magnetic properties different from those of a wrought and homogenized state (Couderchon 1998; Béranger et al. 2009).

Table 2.5 illustrates the mean expansion deviation of an INVAR M93 (Fe-36Ni-0.3Mn-0.03C) measured in two distinct metallurgical states.

Metallurgical state	$\alpha_{20^{\circ}\text{C}-100^{\circ}\text{C}} (^{\circ}\text{C}^{-1})$
Hot rolled sheet (25.4mm)	1.8×10^{-6}
Wall built by additive manufacturing (Arc-wire CMT)	2.3×10^{-6}

Table 2.5. Average dilation α between 20°C and 100°C of INVAR M93

2.2.4.2.5. AM of alloys devoid of refractory elements (Fe-Ni-Cr-Cu-Mn alloys)

Research into productivity is one of the key areas of research in WAAM. Thus, a volume flow $D_v \sim 500 \text{ cm}^3/\text{h}$ is targeted, which corresponds to the mass flow rate of 4 kg/h for a density of 8 g/cm³. However, it is necessary to differentiate the productivity in arcing time (deposition rate) from the total productivity including the cooling times. It is important to take these two phases into account in

the manufacturing part, because the decrease in the pause time increases the manufacturing yields, but can generate the collapse of the walls following the increase in interpass temperatures.



a)



b)

Figure 2.16. Arc-wire additive manufacturing (Fronius CMT). Deposition rate ~ 4.7 kg/h, pause time = 15 s (Aperam results). a) With wire A, alloy Fe-36Ni, the wall collapses from the sixth pass (~ 600 °C); b) with wire B, alloy Fe-36Ni-X, the wall does not collapse until the 15th pass (~ 800 °C)

Figure 2.16 illustrates the construction of walls with MIG Fronius CMT, using two different wires (wire A and wire B), and Fe-Ni alloy, with a pause time deliberately limited to 15 s between passes. With wire A, the wall sinks from 600°C (temperature of pass N-1) while it only collapses above 800°C (temperature of pass N-1) with wire B. The collapse temperature was determined using a FLIR A655sc thermal imager (0.5 im/s).

The presented example highlights that the addition of a refractory alloy element (element X) pushes the limits of the arc-wire process (technological requirement). It is also necessary that this element does not degrade the functional properties (Aperam study).

2.2.4.2.6. AM of aluminum alloy parts

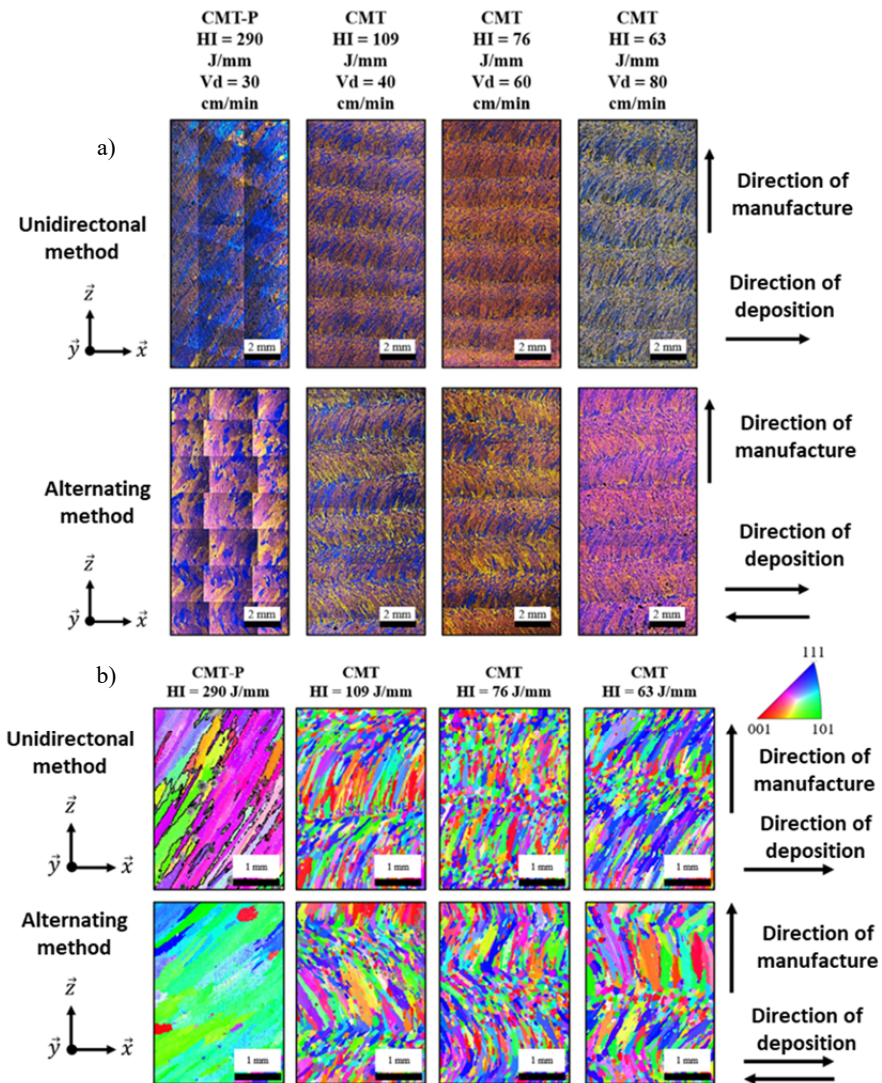
The best-performing aluminum alloys in terms of mechanical properties are those of the age hardening alloy series: 2000 (aluminum–copper), 6000 (aluminum–magnesium–silicon) and 7000 (aluminum–zinc). The wires available for metal AM are few in these grades of aluminum alloys. In fact, in welding, the only grades available are in the 1000, 4000 and 5000 series. These choices of filler metals, despite a greater diversity of base metals, are mainly based on metallurgical issues: the appearance of hot cracking in homogeneous welded parts or in AM with age hardening alloys. Welding, and therefore metal AM by wire fusion with these grades, leads to parts with weak mechanical properties compared to parts made from age hardening alloys of the 2000, 6000 and 7000 series. Studies on homogeneous welding (Benoit et al. 2013) have made it possible to obtain welds which, after heat treatment, present superior properties compared to rolled material of the same grade. Recent work on AM in alloy 6061 has shown the possibilities of producing parts without defects (porosities or cracks) and with superior properties to sheet metal of the same grade (Doumenc 2021).

Figure 2.17 shows microstructures obtained with different energies and different deposition strategies (layers deposited alternately back and forth or unidirectionally, always in the same direction).

The deposition energy can have an impact, but the deposition method has a particularly strong influence on the microstructures obtained: large elongated grains for unidirectional deposition in the same direction and “chevron” grains in the case of alternating deposition.

Following the manufacture of the various walls, non-destructive tests were carried out, thus showing the good material soundness (no porosities and no cracks detected), and mechanical tests (microhardness mapping and tensile tests) were carried out in order to determine the parts’ properties. The deposition energy and the transfer mode have no real impact. On the other hand, the deposition method acting on the microstructure does affect the mechanical properties. Figure 2.18 shows the changes in hardness following aging and/or precipitation treatments. Higher microhardnesses are observed in the case of the alternate deposition method, regardless of the post-deposition treatment. It should be noted that the average micro-hardness values obtained following the T6 treatment are higher than the experimental values obtained on 6061 T6 sheets and the data from the ASM Metal Handbook: 107 Hv. Moreover, following the tensile tests, the mechanical properties obtained are slightly below (5%) the theoretical values obtained on rolled sheets. After fracture surface

analysis by scanning electron microscopy, it is possible to attribute this behavior to the presence of an undetected microporosity during non-destructive tests.



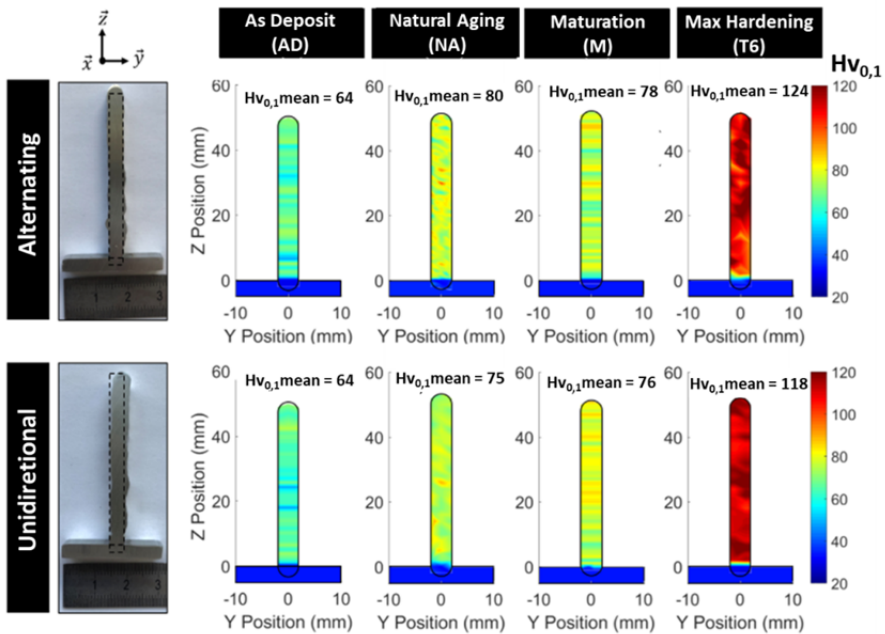


Figure 2.18. Evolution of mechanical properties ($HV_{0.1}$ microhardness mapping) depending on the deposition method (unidirectional or alternating) on the wall being manufactured or following precipitation aging or heat treatment. For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

Finally, we cannot just complete this section without mentioning the requirements related to the formats of materials dedicated to metal AM (powders, wires, strips). The market will have to take into account the fact that the analytical specifications will be dependent on the material formats. For example, the oxygen content of a metal powder made of a UNS N06625 or N07718 alloy is higher than that of a welded wire: less than 10 ppm for a solder wire and more than 100 ppm for a powder with sizes between 20 μm and 45 μm . Likewise, certain grades accessible to powder metallurgy (superalloys) will remain impossible to produce industrially in the form of wires.

According to Aperam, the development plan for new alloys for the manufacture of wires in AM is to build a partnership with academics, machine manufacturers and industrialists who market the final part in order to set up wire

specifications. At this point, the process is structured into six steps (Reydet and Jouvenceau 2019):

- searching for optimized chemical compositions by means of thermodynamic calculations, non-equilibrium kinetic calculations (ThermoCalc, JMatPro) or computer-aided alloy design (artificial intelligence, genetic algorithms, etc.) (Menou et al. 2016);

- mini-development of innovative chemical compositions using vacuum induction furnaces;

- mini-hot processing of ingots by high temperature rolling, and production of study materials;

- characterization of properties, either directly on the ingots (solidification structures) or after hot transformation in the form of beams. The aim is to spot trends:

- evaluation of resistance to hot cracking (Varestraint tests or PVR test),
- measurement of mechanical properties (R_p , R_m , A%, KCV, HV, etc.),
- characterization of physical properties (dilatometry, magnetic properties),
- microstructural characterizations to make the link with the properties (SEM, EBSD, TEM);

- choice of chemical composition, followed by production of industrial castings and industrial wires;

- evaluation of the “alloy solution” using an AM robotic cell.

2.2.5. Summary

This section provided a rapid inventory of technologies based on wire as input material, with particular emphasis on the most widespread arc-wire process (WAAM). Like powders, good filler wire quality, or even dedicated preparation for AM, results in better final properties in the deposits. Wire technologies allow a deposition speed greater than that of powder deposits, which makes them suitable for the manufacture of large parts. Few parts are made in this way. Regarding the three technologies that are distinguished mainly by the energy source (electric arc, laser or electron beam), each of them have their own specificities, which makes transpositions difficult, even for the same material. Many materials (steels, Al base, Ti base, Ni base) are available in the form of wires, but at the present time there is

no database on the respective wire qualities from different manufacturers and therefore on the risks of reducing mechanical properties.

2.3. References

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3

The Physics of Metal Additive Manufacturing Processes

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Metal additive manufacturing processes use physical principles often inspired by welding that directly deal with the absorption of incident power (electric arc, laser or electron beam), temperature fields and fluid velocity in the fusion zones, with particular boundary conditions due to the presence of powder or filler wire. These processes are also subject to instability-generating defects in 3D parts. This chapter presents a summary of the main physical phenomena involved in these processes.

3.1. The energy–powder–fusion zone interaction in additive laser fusion processes

3.1.1. Introduction

The optimized implementation of additive manufacturing processes using laser fusion requires good control of the conditions of absorption of electromagnetic radiation with the powder and the fusion zone and the formation of hydrodynamic instabilities in liquid metal baths. This section recalls the main physical concepts involved in additive manufacturing by laser fusion of metal parts, and in particular for laser powder bed fusion (L-PBF) and laser metal deposition (LMD/DED).

Section 3.1 written by Patrice PEYRE and Rémy FABBRO.

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coordinated by Patrice PEYRE and Éric CHARKALUK. © ISTE Ltd 2022.

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3.1.2. Reminder of the essential physical variables

The main physical principles involved in additive laser fusion manufacturing processes (Figure 3.1) are as follows:

- i) the absorption of radiation resulting in the overall interaction efficiency;
- ii) heat conduction that defines the contours of fusion isotherms;
- iii) the hydrodynamics of the fusion zones on which depends the bead stability;

iv) rapid solidification (see section 1.1 of Volume 2) of the molten metal that leads to the formation of metallurgical grains, followed by an annealing heat treatment associated with the accumulation of layers of material.

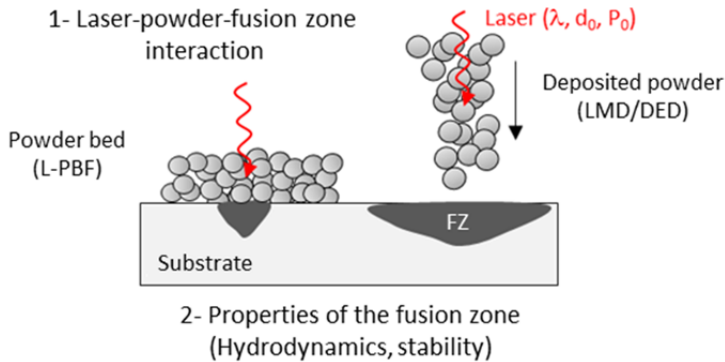


Figure 3.1. Parameters involved during the laser/powder/substrate interaction

From the first-order parameters of laser fusion processes (power absorbed $P_{\text{abs}} = A \cdot P_0$ (W), sweep speed V_0 ($\text{m} \cdot \text{s}^{-1}$), laser diameter D_0 (m)), we can define:

- the surface power density ($\text{W} \cdot \text{m}^{-2}$): $I_0 = P$ (W)/ S_0 (m^2), which is commonly used, with the laser–matter interaction time τ (s), to describe the state of matter (solid, liquid, vapor, plasma) in typical laser processes (Laurens et al. 1996);

- the volume energy density ($\text{J} \cdot \text{m}^{-3}$), which has practical utility and can be described with (at least) two different formulations:

- E_v or DEV ($\text{J} \cdot \text{m}^{-3}$) = I_0 ($\text{W} \cdot \text{m}^{-2}$)/ V_0 ($\text{m} \cdot \text{s}^{-1}$) essentially reflects the size of the fusion zones, and in particular the depth of the fusion beads, which it affects linearly according to analytical welding models (Fabbro 2019),

- E_v^G or EVF ($\text{J}\cdot\text{m}^{-3}$) = P_0 (W)/(V_0 ($\text{m}\cdot\text{s}^{-1}$), Δh (m), e_v (mm)) with Δh = layer height and e_v = hatch is mainly used in powder bed fusion and reflects the amount of energy stored per volume of material manufactured (EVM) (Gunenthiram et al. 2018). As such, it also conditions the microstructure and the residual stresses–deformations within the parts, which are directly related to the thermal gradient G ($\text{K}\cdot\text{m}^{-1}$).

3.1.3. Radiation absorption and heat transfer: different interaction regimes for different processes

3.1.3.1. Laser–material interaction

Laser radiation consists of pulsating electromagnetic waves ω whose propagation and reflection–transmission properties at the interface between air and the solid surface can be studied by solving Maxwell's equations. In the case of a metal, the attenuation of the radiation at depth is calculated according to the Beer–Lambert law (equation [3.1]) with an absorption coefficient α (m^{-1}). The corresponding skin thicknesses ($\sim 1/\alpha$), which increase linearly with wavelength, are a few nanometers for most metals and $\lambda \sim 1.06 \mu\text{m}$ (usual wavelength in AM). Overall, when α is high, the radiation only penetrates into the material a little and the reflectivity R is high:

$$I = I_0 \exp(-\alpha z) \quad [3.1]$$

where α (m^{-1}) = absorption coefficient, I_0 = incident laser intensity ($\text{W}\cdot\text{m}^{-2}$) and z (m) = depth.

Physically, the absorption of laser (monochromatic) photons in metals occurs through the excitation of free electrons located in the conduction band which, when de-excited, heat the crystal lattice (Laurens et al. 1996). The reflectivity of the medium can be expressed as a function of the electrical conductivity σ_0 and the frequency ω of the incident wave, therefore of the wavelength of the laser ($\omega = 2\pi C_0/\lambda$), using the Hagen–Rubens relation:

$$R = 1 - 2 \sqrt{\frac{2 \omega \epsilon_0}{\sigma_0}} \quad [3.2]$$

where ϵ_0 = permittivity of the vacuum ($8.854 \times 10^{-12} \text{ A}^2\cdot\text{s}^4\cdot\text{kg}^{-1}\cdot\text{m}^{-3}$) and σ_0 ($\text{S}\cdot\text{m}^{-1}$) = electrical conductivity.

Metals that are very electrically and thermally conductive (Al, Cu, Ag) therefore have a higher reflectivity (>90%) than transition metals (Fe, Ni, Co, Ti), the latter having absorptivities A ($A = 1 - R$) between 25% and 50% for near-infrared.

The state of the material with which the laser interacts varies according to the process (L-PBF or DED) and the power density considered, which has a direct effect on the values of A . Thus, the transition to the liquid state causes an absorptivity increase of +2% (Ti) to +40% (Ni) (Bretonnet 2005).

The L-PBF process implements a laser–material interaction at I_0 or DEV_0 values approximately 10 times higher than those of DED under normal melting conditions (around $I_0 \approx 10^{4.5} \text{ W}\cdot\text{cm}^{-2}$ in DED and $\approx 10^6 \text{ W}\cdot\text{cm}^{-2}$ in PBF). The associated interaction regime then often exceeds the vaporization threshold in PBF, leading to the formation of a vapor capillary (keyhole), well-known in laser welding. Conversely, DED fusion is less energetic and does not vaporize the liquid metal, with the exception of elements with a low vaporization point (the case of aluminum in Ti-6Al-4V). We therefore stick to the conduction welding regime and the cross-section geometry of the fusion zones (FZs) is quasi-elliptical.

The keyhole melting regime begins when the surface of the liquid is deformed by the recoil pressure imposed by the vertical expansion of the metal vapor (Figure 3.2). This recoil pressure P_{rec} can be estimated via its dependence on the surface temperature using the Clausius–Clapeyron law (equation [3.3]), giving the saturating vapor pressure P_{sat} , corrected by a recondensation factor ($k \approx 0.55$ at 1). The corresponding pressure values, strongly dependent on the materials, are around 0.1 bar to a few bars for the temperatures reached with PBF processes. A dynamic equilibrium is then created between the relative displacement of the laser with respect to the substrate (or to the powder) and the vertical deformation of the liquid (by a “piston” effect). The result is a rather deep vapor capillary, the front of which is inclined θ with respect to the vertical. One can reasonably define a keyhole formation threshold when the aspect ratio H_{FZ}/D_0 (molten metal depth/laser diameter) ≈ 1 , for which the leading edge of the capillary is tilted 45° . For $H_{\text{FZ}}/D_0 > 1$, the laser beam, which is reflected optically downwards on the walls of the capillary, thus contributes to greater energy deposition and therefore to greater evaporation, which results in increased penetration into the material:

$$P_{\text{rec}}(T) = k P_{\text{sat}}(T) = k P_0 \exp\left[\frac{ML_v}{k_b T_v} \left(1 - \frac{T_v}{T}\right)\right] \text{ with } 0.55 < k < 1 \quad [3.3]$$

where P_{rec} (Pa) = recoil pressure, P_{sat} (Pa) = saturating vapor pressure, M (kg) = atomic mass, L_v ($\text{J}\cdot\text{kg}^{-1}$) = enthalpy of vaporization, T_v (K) = vaporization temperature and $k_b = 1.38 \times 10^{-23} \text{ J}\cdot\text{K}^{-1}$ = Boltzmann constant.

To go further, a prediction of the overall absorptivity of the capillary resulting from the multireflection phenomena on the walls of the keyhole is possible via a blackbody geometric model (Metel et al. 2018).

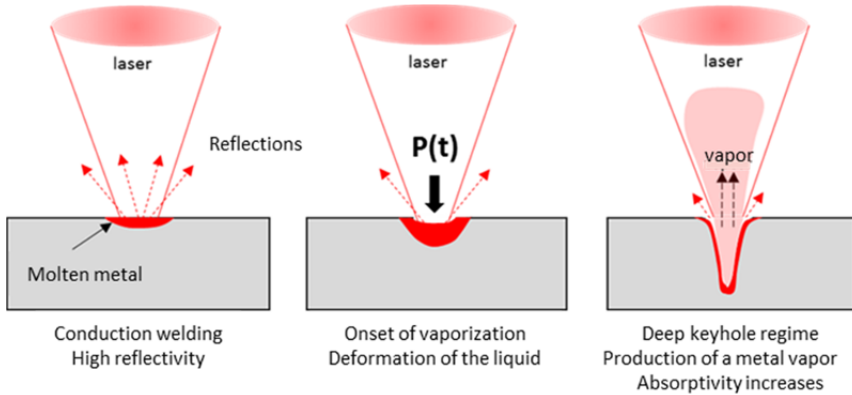


Figure 3.2. Laser fusion regimes (from conduction to keyhole).
The deformation of the liquid surface begins when
 P_{rec} is greater than ambient pressure

3.1.3.2. Laser absorption in DED

During additive manufacturing by DED with a coaxial powder nozzle (the most common case), the laser first interacts with the powder cloud projected over an interaction distance d_i before being absorbed by the substrate (the previous layer) (Figure 3.3). The absorption of incident energy therefore takes place partly in the powder grains, thus leading to heating of the latter, and partly in the material already manufactured. The proportion of power absorbed logically increases with the mass flow D_m ($\text{g}\cdot\text{min}^{-1}$) on which the powder volume density depends N (m^{-3}) and with the laser/deposited powder interaction time $= l_w / V_p$ (with V_p = velocity of powder grains). Different models, including that of Qi et al. (2006), thus make it possible to estimate the laser intensity I' transmitted through the projected powder and heating ΔT_p of powder grains (equations [3.4] and [3.5]). In the usual configurations, a high proportion of the power (~ 70 – 90%) is absorbed by the substrate, thus creating a fusion zone fed by projected particles still in the solid state. By increasing l_w , or D_m ,

it is possible to switch from a fusion regime for the grains of powder and to tend toward a bonding regime for the particles on a solid substrate:

$$I'(r, z) = I_0(r) \cdot \exp(-S_p N l_w) \quad [3.4]$$

$$\Delta T_p = A \frac{z_0}{m_p c_p V_p} S_p (I_0 - I'(r, z)) \quad [3.5]$$

where $I'(r, z)$ ($\text{W} \cdot \text{m}^{-2}$) = transmitted intensity, S_p (m^2) = cross-section of the powder grain, N (m^{-3}) = volume density of powder grains, l_w (m) = work distance = powder–laser interaction distance, m_p (kg) = mass of a grain of powder, V_p ($\text{m} \cdot \text{s}^{-1}$) = powder grain velocity and C_p ($\text{J} \cdot \text{kg}^{-1} \cdot \text{K}^{-1}$) = specific heat.

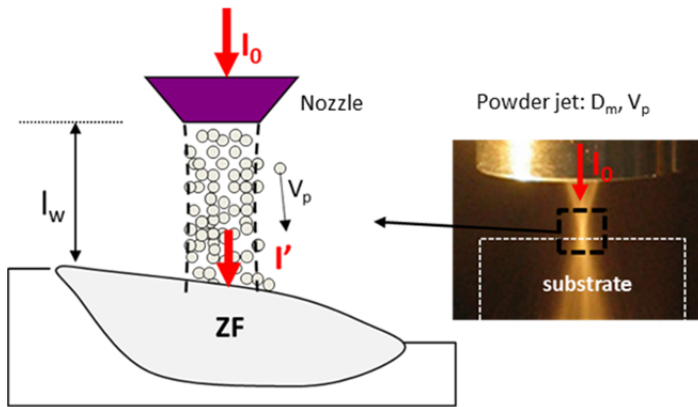


Figure 3.3. Powder jet/laser/substrate interaction in DED

3.1.3.3. Laser absorption in L-PBF

One of the recurring questions is whether, in a quasi-stationary L-PBF melting regime, the laser radiation is absorbed in the powder bed or in the molten zone. Regarding the powder bed, Gusarov and Smurov (2010) have proposed models to describe its absorptivity from the reflectivity R_{grain} of the grains (equation [3.6]). We find the geometric trapping effect of the radiation via multireflections on the powder grains, which increase the overall absorptivity up to 70–80%, or double that of the grain (Figure 3.4):

$$A = \frac{3\sqrt{1-R_{\text{grain}}}}{1+2\sqrt{1-R_{\text{grain}}}} \quad [3.6]$$

where R_{grain} = reflectivity of the solid powder grain.

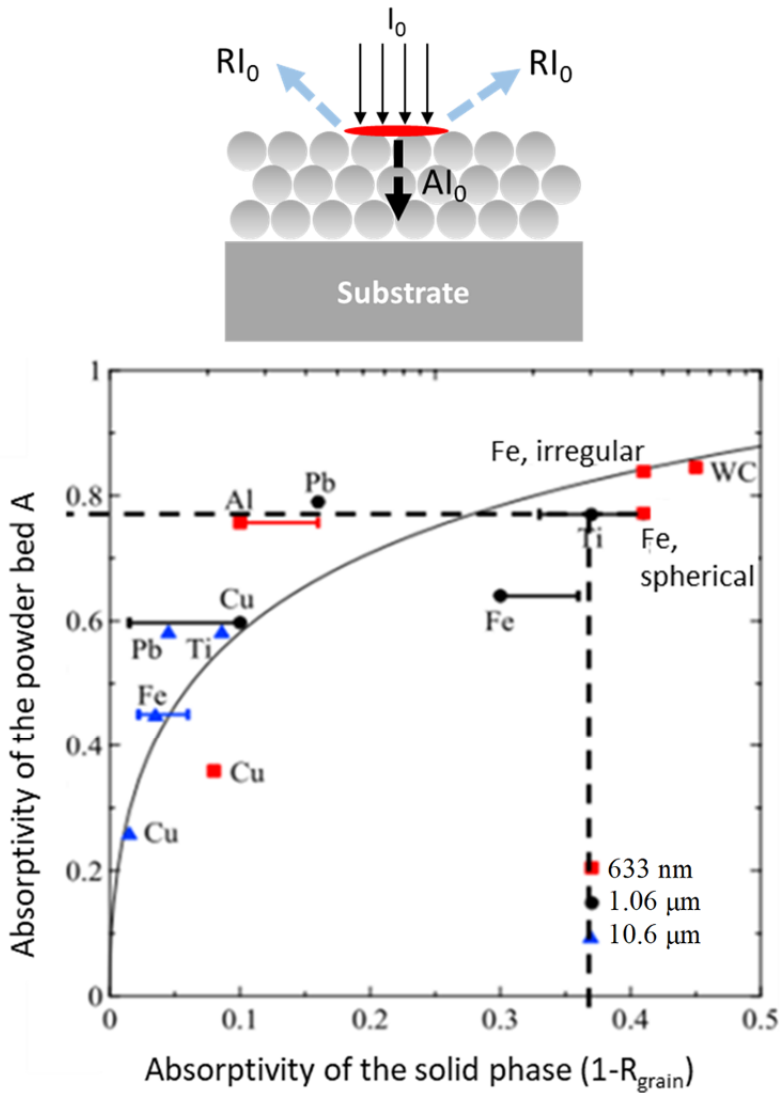


Figure 3.4. Absorptivity of a powder bed according to Gusarov and Smurov (2010). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

Recent analyses of the laser–fusion zone interaction in L-PBF (Gunenthiram et al. 2018) show, however, that the laser beam is most often absorbed in the fusion zone surrounding the laser spot, which supports the theory that most of the radiation is absorbed in the liquid. In certain types of materials (very thermally conductive such as aluminum or copper), the small size of the fusion zones can however lead to a mixed absorption regime (with the Gaussian laser beam positioned both in the fusion zone and the powder bed upstream of the FZ).

However, with the high power and energy densities involved in L-PBF, which favor a keyhole regime, the trapping of the laser radiation still occurs via multireflection, but this time on the liquid walls of the vapor capillary. This geometric effect is responsible for the great fusion depths – often greater than three times the diameter of the L-PBF beam. The overall absorptivity of radiation A therefore increases at high energy and we obtain the average curve in Figure 3.5, as confirmed by Trapp et al. (2017).

Furthermore, the stripping mechanisms generated at the edge of the beads can cause the laser to no longer interact directly with the powder bed, but with an area stripped by the passage of the previous bead. This phenomenon of indirect melting by denudation/attraction of the powder grains toward the fusion zone is detailed in Figure 3.15, later in the chapter. It is all the more marked when the hatch is small, so that the beads can be confined to the bare zone of the previous bead.

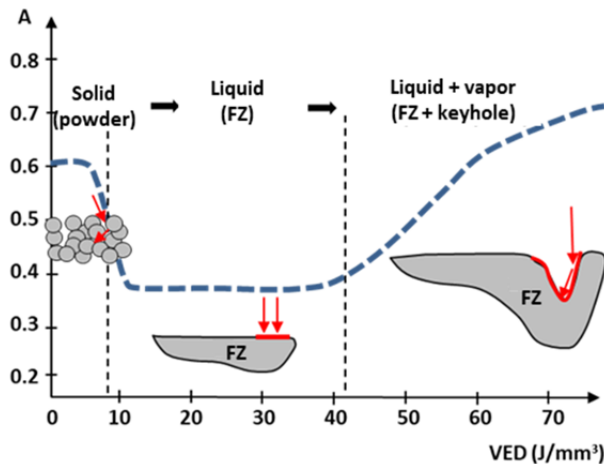
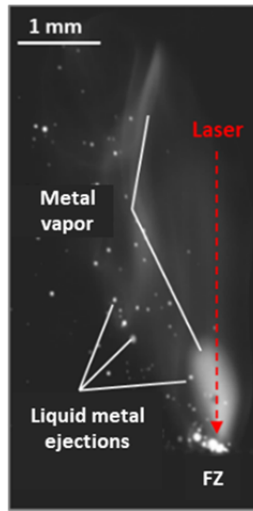


Figure 3.5. Different types of laser absorption in PBF: on the solid powder bed, on the FZ surface, on the keyhole walls. From Trapp et al. (2017). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip



b)

Figure 3.6. a) Front view of the laser/powder/fusion zone/metal vapor interaction in L-PBF according to Gunenthiram (2018). b) Vapor release in L-PBF according to Gunenthiram (2018)

In all cases, a hot concentrated metal vapor limits the coupling with the powder and the FZ by attenuating (by a few percent) the beam transmitted according to the Beer–Lambert law (equation [3.7]). The optimization of gas flows in the PBF enclosures, and in particular that of the cross-jet close to the build plate, can then clean the optical path of the laser:

$$P_{att} = P_0 [1 - \exp(-(Q_{Diff} + Q_{abs})\pi r^2 N z)]$$

$$Q_{Diff} = \frac{8}{3} \left(\frac{2\pi r}{\lambda} \right)^4 \left| \frac{n_r^2 - 1}{n_r^2 + 2} \right|^2 \quad \text{and} \quad Q_{Abs} = -\frac{8\pi r}{\lambda} \text{Im} \left\{ \frac{n_r^2 - 1}{n_r^2 + 2} \right\} \quad [3.7]$$

where P_{att} (W) = attenuated power, P_0 (W) = incident power, Q_{Diff} = optical diffusion coefficient, Q_{Diff} = optical absorption coefficient, r (m) = laser radius, N (m^{-3}) = particule density, z (m) = interaction height, n_r = complex refraction index and λ (m) = laser wavelength.

3.1.3.5. Estimating the dimensions of fusion zones (L-PBF and DED-LMD)

There are different analytical models that estimate the dimensions of fusion zones by laser welding (and in L-PBF) as a function of the process parameters

(P_{abs} , V_0 , D_0) and thermophysical properties of the material (density ρ , specific heat C_p , heat conductivity K). These models establish linear dependencies between the geometric variables of the beads (length L_{ZF} , width l_{ZF} , depth H_{ZF}), standardized by the laser diameter D or depth of heat diffusion δ ($=\sqrt{a\tau}$), where $\tau = D_0/V_0$ = laser–material interaction time, and laser energy density ΔH_0 ($\approx AP_0/(D_0^2 V_0)$), standardized by enthalpy ΔH_F at the onset of melting ($=\rho C_p(T_F - T_0)$) (King et al. 2014) or at the onset of vaporization ΔH_v ($=\rho C_p(T_v - T_0)$) (Fabbro 2019). Such models have been experimentally validated (Figure 3.7) taking into account the thermophysical properties (K , ρ , C_p) in the liquid state. They thus allow parametric optimization assistance by limiting the experimental designs based on a large number of single fusion beads:

$$Y = \frac{H_{ZF}}{\delta} = \beta \frac{\Delta H}{\Delta H_F} \quad \text{with} \quad \beta = \frac{\pi^{1/2}}{2^{3/4}} \frac{T_F}{m(T_v - T_0)} \quad [3.8]$$

$$R = \frac{H_{ZF}}{D} = \frac{\left(\frac{AP_0}{VD_0^2}\right)}{0.5 m \rho C_p (T_v - T_0)} = \frac{\Delta H_0}{\Delta H_v} \quad [3.9]$$

where T_F (K) = fusion temperature, T_v (K) = vaporization temperature, T_0 = initial temperature, m = parameter of a linear function $g(\text{Pe}) = m \text{Pe} + n$ of the Péclet number ($\text{Pe} = V_0 D_0 / 2a$), with $(m, n) = (2.5, 3)$ for $0.5 < \text{Pe} < 10$.

In the particular case of the DED-LMD process, the laser melting regime is closer to a purely conductive regime (with limited vaporization), due to lower power densities (from 50 to 200 $\text{kW} \cdot \text{cm}^{-2}$) than those in L-PBF ($>200 \text{ kW} \cdot \text{cm}^{-2}$). The fusion widths and depths can then be estimated with solid thermal calculations, or by using analytical models such as the Rosenthal model, which considers a point source deposit. On the other hand, the height of material produced, as well as the interaction efficiency, are directly related to the accumulation time of the projected material (D_{powder}/V), and at the interaction surface between the powder jet and the fusion zone (Figure 3.8). An analytical formula of the evolution of the manufactured heights then gives:

$$\Delta h = \frac{1}{\rho V l_{FZ}} \iint_{FZ} D_m^*(x, y) dx dy \quad [3.10]$$

where $D_m^*(x, y)$ is the distribution of the surface mass flow rate of the powder ($\text{kg} \cdot \text{s}^{-1} \cdot \text{m}^{-2}$).

The influence of the powder spatial distribution $D_m^*(x, y)$, variable from one nozzle to another and often close to a Gaussian function, is then evident with regard to the geometries of the beads, but also the powder/fusion zone interaction yields, easily estimated using the same type of approach.

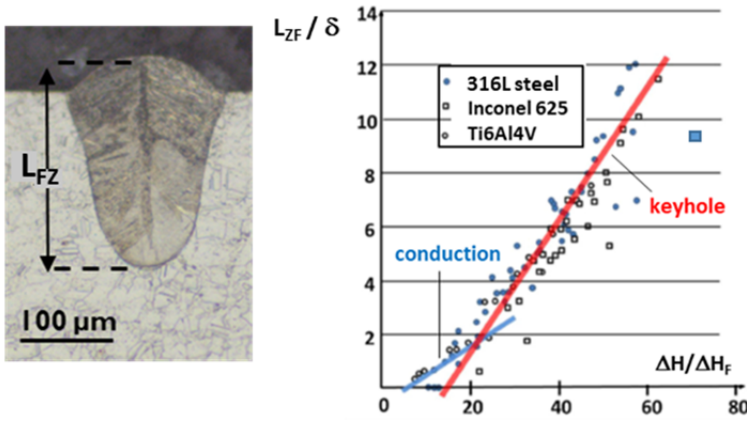


Figure 3.7. Standardized melted depths L_{FZ}/δ in PBF for three materials with similar Pe values ($2 < Pe < 5$). The slope of the regression line (0.25) is close to the analytical calculation of β (0.27) via equation [3.9]. From Fabbro (2019). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

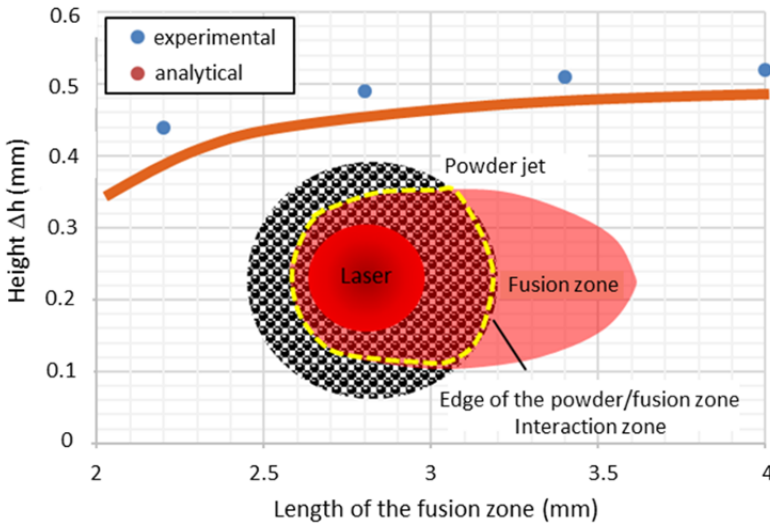


Figure 3.8. Calculation of heights manufactured in DED-LMD. Diagram and example of comparison of experimental results and the analytical approach (Ferreira 2021). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

3.1.4. Local thermal cycles: influence of boundary conditions

The L-PBF process implements a faster melting-solidification process than the DED process (solidification rate $\sim \text{m} \cdot \text{s}^{-1}$ vs. $\text{mm} \cdot \text{s}^{-1}$). In both cases, at the mesoscopic scale (the fusion bead and its immediate environment), the thermal cycle seen by the material conditions the solidification structure (columnar, equiaxed, fine) while at a more macroscopic scale, it is the temperature differences ΔT and/or the average temperatures reached which define the deformations and instantaneous and residual thermal stresses (σ_{th}):

$$\sigma_{th} = \frac{E \alpha_{th} \Delta T}{(1-\nu)} \quad [3.11]$$

where ΔT (K) = temperature difference, α_{th} = coefficient of thermal expansion (K^{-1}), ν = Poisson coefficient and E = elasticity modulus (Pa).

What distinguishes the overall thermals induced by the DED and L-PBF processes is the average temperature reached, which is generally higher in DED. The cooling time between layers (longer in L-PBF) and the average energy injected per unit of volume manufactured (E_v^G or EVF ($\text{J} \cdot \text{m}^{-3}$) = $P_0/(H \cdot \Delta h \cdot V_0)$) is greater in DED, making it possible to explain this result.

In addition to the intrinsic parameters of the laser-matter interaction (P_0 , V_0 , D_0), the manufacturing process is a major issue in controlling the thermal-mechanical evolution of a part, just like its geometry. Thus, in DED, a “zig-zag” strategy, more efficient in terms of relative movement of the powder-part nozzle, promotes the concentration of heat at the edge of the part and promotes the accumulation of powder in FZ. The result is a typical geometry with elevations at both ends of the fabricated wall. Using a time pause between each fusion line and switching to a one-way laser strategy are stabilization solutions. The influence of laser methods is even more visible in L-PBF. So, for a circular part, a concentric strategy sweeping from the outside to the inside of a part will concentrate the heat in its center. Likewise, a zig-zag scan strategy, oriented perpendicular to the sharp corner of a part, will have the same overheating effect (Figure 3.9), reducing the cooling time between two successive laser lines. This phenomenon is amplified by the thermal boundary condition of areas with a small radius of curvature, which limits heat dissipation.

In all cases, as explained in Chapter 4, solving the heat equation provides a good idea of the local thermal cycles seen by the material and the dimensions of the fusion zones formed during laser AM processes from a surface source deposit I_0 (W/m^2) = $f(x,y)$ and the thermophysical properties of the materials considered.

Numerically, the management of the supply of material remains a key factor (see Chapter 4 of this volume), just like the description of the equivalent properties of a powder bed consisting of at least 40% gas:

$$\rho C_p \left(\frac{\partial T}{\partial t} + \vec{u} \cdot \nabla T \right) = \nabla \cdot (K \nabla T) + S_v \quad [3.12]$$

where S_v ($\text{W} \cdot \text{m}^{-3}$) = volume source term (= 0 in case of surface laser deposition).

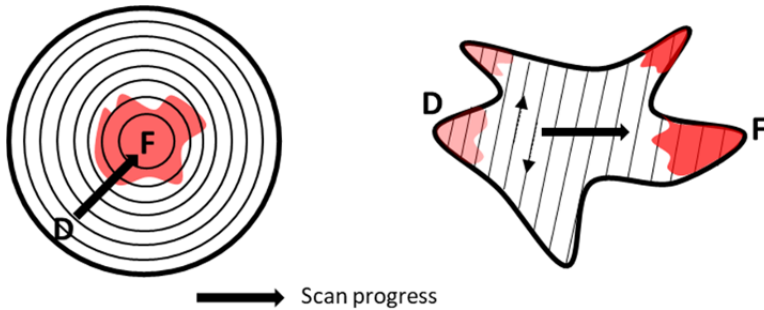


Figure 3.9. Influence of L-PBF manufacturing strategies: configurations to be avoided (in red: areas of overheating linked to the shortening of the laser lines).
For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

3.1.5. Hydrodynamics of fusion zones and associated faults

Additive manufacturing by laser fusion is equivalent to moving a molten zone in a three-dimensional space, delimiting the layer being manufactured. The dimensions of these FZs are between 0.1 and 0.5 mm in L-PBF and 10 times larger in DED. At the center of these fusion zones, different convective currents exist, often at high speeds ($\approx \text{m/s}$), which tend to homogenize both the FZ temperatures and the chemical compositions. The latter are however likely to vary at the local scale from microsegregation during solidification (e.g. enrichment of interdendrite or intercellular spaces in the alloying element).

3.1.5.1. Origin of convection movements

As in all laser fusion processes, thermocapillary (so-called Marangoni) convection is one of the main drivers of the hydrodynamics of fusion zones for laser additive manufacturing processes. It is directly caused (Figure 3.10) by the surface tension gradients γ of the liquid metal, themselves linked to temperature gradients G ($\text{K} \cdot \text{m}^{-1}$) in the FZ. The latter are particularly high in DED ($\approx 10^6 \text{ K} \cdot \text{m}^{-1}$) and more so in L-PBF ($\approx 10^7 \text{ K} \cdot \text{m}^{-1}$). This results in centrifugal movements (the most common

case being where the Marangoni coefficients are negative) that widen the beads or centripets (positive Marangoni coefficients), which make them fall.

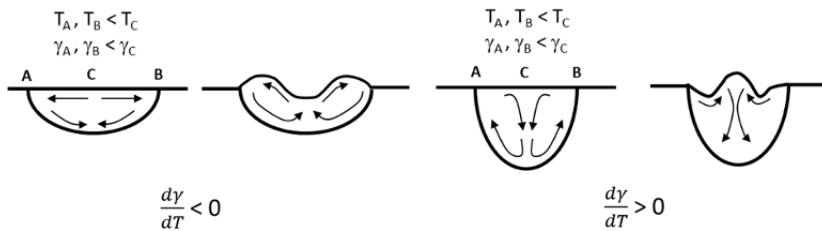


Figure 3.10. Principle and origin of Marangoni convection

When the surface of the fusion zone is deformed by the recoil pressure due to vaporization, it results in a second mode of convection directly induced by the surface deformation dynamics (piston effect) and by the formation of the keyhole regime. Recoil pressure P_{rec} exerted on the front of the capillary generates a fluid flow (Figure 3.11) at high speed toward the rear of the beads with two components: (1) shearing between the liquid and the metallic vapor at the rear of the capillary (speed V_1) and (2) the Venturi effect – acceleration of the liquid at the FZ surface due to constriction between the solid side of the bead and the vertically deformed liquid zone (speed V_2).

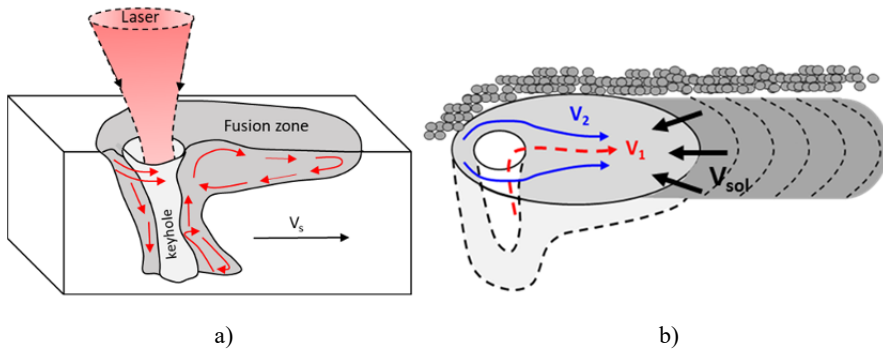


Figure 3.11. Convection movements induced by a keyhole regime in conventional laser welding through the outlet a) and FZs in L-PBF b). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

The dynamic pressure ($0.5 \rho V_F^2$) exerted on the capillary by the moving liquid, combined with surface tensions (γ/r), then opposes the dynamic recoil pressure that

keeps the capillary open. The effects of other sources of convection (buoyancy) can be considered negligible compared to the two contributions mentioned above.

3.1.5.2. Defect formation: the case of the laser powder bed fusion process

The complex fluid movements described above can be the source of hydrodynamic instabilities such as humping: the bead intermittently empties backwards due to excessive fluid speeds in the FZ, resulting in the formation of a succession of humps and valleys. This phenomenon, which is fairly regular in PBF at high speed and power, can hinder the densification of the material by limiting the homogeneous overlap between the beads. More classically, the balling phenomenon (Figure 3.12) is generally observed for insufficient energy inputs and is associated with insufficient melting of the previous layer (or lack of dilution).

It results in the formation of discontinuous droplets in place of a continuous fusion bead, ensuring a minimization of the free energy of the system. A destabilization–spheroidization threshold for the fusion zone has been proposed by Gusarov and Smurov (2010). Such a criterion, based on the formation of Plateau–Rayleigh instabilities, indicates that for a cylinder of free liquid metal, balling occurs under the effect of variations in local pressures exerted on the liquid ($P = \gamma \left(\frac{1}{R_1} + \frac{1}{R_2} \right)$) for a dimensional ratio $L_{FZ}/l_{FZ} > \pi$ (with L_{FZ} and l_{FZ} = length and width of fusion zone).

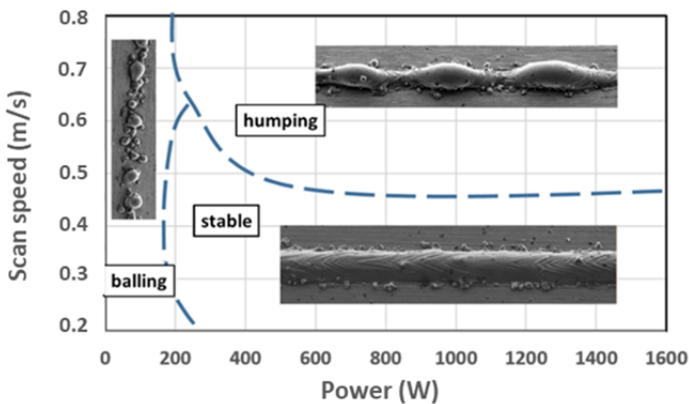


Figure 3.12. Mapping of the domains of fusion stability on 316L steel in L-PBF, with $D = 200 \mu\text{m}$ and $60 \mu\text{m}$ of powder, according to Gunenthiram (2018)

The presence of residual defects (porosities) in L-PBF parts can be explained by the following different factors:

1) Poor optimization of process parameters, which generates insufficient overlap between beads and lack of fusion. Tang et al. (2017) have proposed a very simple geometric approach (Figure 3.13), which defines a densification criterion as a function of dimensional ratios (hatch/bead width = e_v/l_{FZ} and total bead height/layer height = $H_{FZ}/\Delta h$), considering the fusion zones as the sum of two half-ellipses. Such a criterion gives a good idea of the values of Δh and e_v to be used according to the dimensions (H_{ZF} , l_{ZF}) of a single bead:

$$\left(\frac{e_v}{l_{FZ}}\right)^2 + \left(\frac{\Delta h}{H_{FZ}}\right)^2 \leq 1 \quad [3.13]$$

2) Capillary instabilities and closures, which are typically the cause of fusion bead root cavities. This phenomenon is frequent for L-PBF conditions with high volume energy densities, or in areas sensitive to scanner head inertia (start and end of laser vectors, contour/fill transition) subject to acceleration–deceleration phenomena and very low local scanning speeds. The skywriting component (the scanner head performs its acceleration and deceleration phases outside the part) installed on L-PBF machines limits this phenomenon (Mancisidor et al. 2016).

3) The presence of porosities at the center of the powder grains, linked to the atomization process, both in DED and PBF.

4) A competition between the flow of liquid metal by capillary action and rapid solidification. According to Zhou et al. (2015), materials with faster solidification times τ_s (equation [3.14]) mean less time for the capillarity-densification process to be established and make densification more difficult. This is the case with Cu, W and Al alloys (Figure 3.14). Other factors such as the presence of residual O_2 in oxide form, for example aluminum, can play an inertial role and hinder densification:

$$\tau_s = 2 \left(\frac{r_g^2}{3a_{th}} \right) \ln \left(\frac{T - T_0}{T_F - T_0} \right) \quad [3.14]$$

where τ_s (s) = solidification time, a_{th} ($m^2 \cdot s^{-1}$) = thermal diffusivity, r_g (m) = radius of a liquid droplet, T_F (K) = fusion temperature, T_0 (K) = ambient temperature and T (K) = temperature of the droplet.

During L-PBF fusion, two other types of phenomena comparable to defects are also present: (1) the ejection of spatters likely to be redeposited on the powder bed or on the parts during manufacture and (2) denudation of the powder on either side of the fusion beads, which can interfere with the optimization of the inter-bead

overlaps. Although the complex physical mechanisms behind these two phenomena are not fully understood, it seems (Matthews et al. 2016) that a significant contribution comes from the presence of the metal vapor column which, by shearing the surrounding gas, creates a gas recirculation loop that pushes the powder grains toward the fusion zone, causing denudation (Figure 3.15). The beads formed therefore consist both of the powder present in the path of the laser (direct fusion) and that nearby which is attracted laterally toward the fusion zone and contributes to the growth of the material (indirect fusion). Recent numerical and analytical studies (Mayi et al. 2020) have modeled this phenomenon and established lateral fluid velocity thresholds beyond which the dynamic pressures ($1/2 \rho_{\text{gas}} V_{\text{gas}}^2$) are greater than the interparticle interaction forces (see Chapter 4 of this volume).

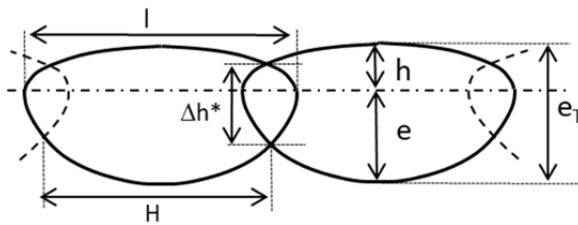


Figure 3.13. Illustration of the geometric densification model of Tang et al. (2017). Densification is optimal when $\Delta h^* > \Delta h$

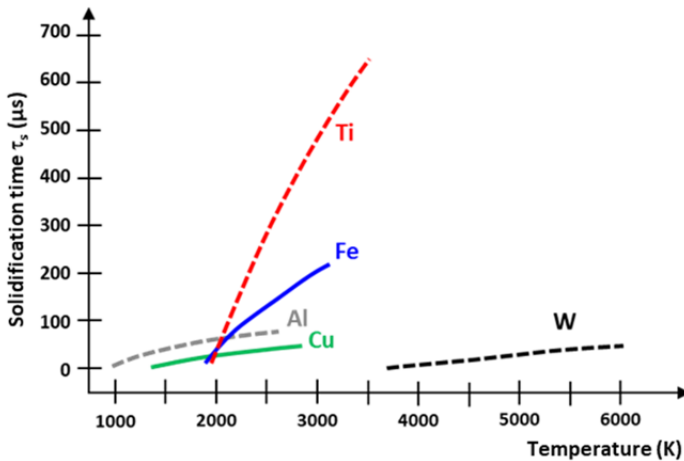


Figure 3.14. Estimate of the solidification time τ_s for different alloys used in L-PBF. The longest τ_s (Ti, Fe) is conducive to densification (according to Zhou et al. (2015)) for a radius of the liquid droplet $r_g = 100 \mu\text{m}$. For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

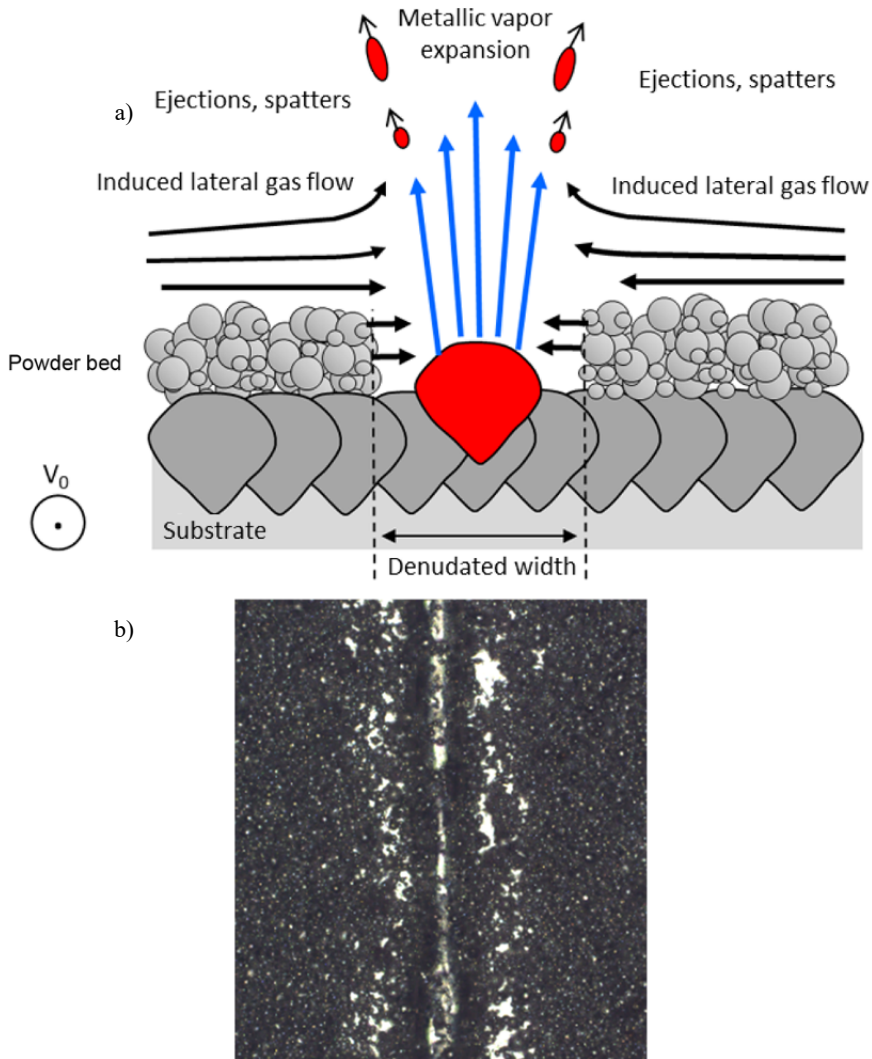


Figure 3.15. L-PBF denudation phenomenon. a) Influence of the metal vapor/shielding gas friction and formation of a lateral flow which pushes the powder grains towards the FZ (adapted from Matthews et al. (2016)). b) Example of denudation around a single L-PBF bead. For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

In the end, these different types of defects are often erased by the successive reflections from the following layers (each bead remelts at least two or three

previous layers). These are implemented with angles of rotation θ , which favor the densification of the material.

Finally, the recent use of dedicated physical experiments, carried out for example by time-resolved X-ray radiography on a synchrotron, makes it possible to provide very valuable information on the physics of the laser-fusion zone and powder-bed interaction. Probing the material during L-PBF fusion makes it possible to visualize very clearly the complex dynamics of the phenomena involved (Figure 3.16), and in particular the instability of the vapor capillary, even if the experimental conditions used (one single bead, powder bed reduced in width) deviate from the actual conditions used with the machine.

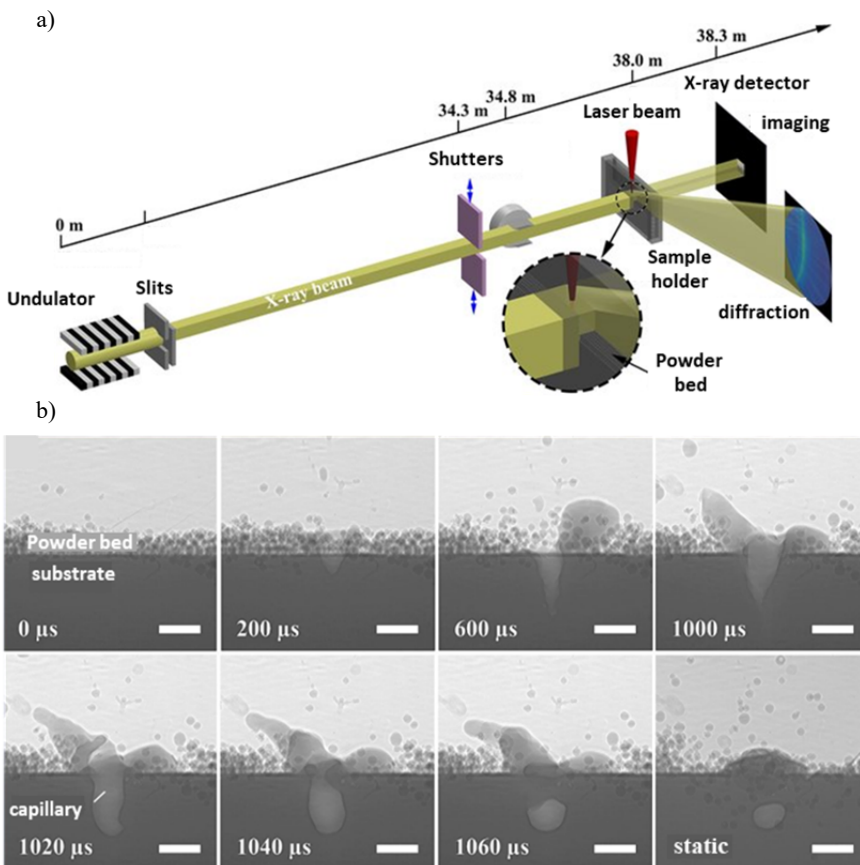


Figure 3.16. High-speed X-ray image of an L-PBF fusion bead. a) Principle of the experiment. b) Successive images of the laser–powder and fusion zone interaction (X-ray imaging at 50 kHz, $P_0 = 520$ W, $D_0 = 220$ μm). From Zhao et al. (2017)

3.1.6. Partial conclusion regarding the physics of laser additive manufacturing processes

Laser additive manufacturing processes have complex physics, often derived from laser welding, around the laser–powder–molten zone interaction and gaseous environment. The radiation absorption conditions and the stability of the fusion zones formed from the initial powder are the main factors to be controlled to ensure the quality of the manufactured parts, these two criteria being largely dependent on the thermophysical properties of the materials considered and on the optimization of the operating parameters used.

3.2. The physics of the E-PBF process

3.2.1. Introduction

The additive manufacturing process using an electron beam for the fusion of a powder bed is an interesting alternative to the laser in terms of absorption of the energy transmitted to the powder. Once the material has melted, the liquid zone is comparable to that obtained by a laser, the site of hydrodynamic instabilities in liquid metal baths. This section recalls the main physical concepts involved in additive manufacturing by electron powder bed fusion (E-PBF) or electron beam melting (EBM) of metal parts, and in particular for powder bed fusion processes.

3.2.2. Reminder of the essential physical variables characteristic of the electron–matter interaction

The main physical phenomena involved in additive manufacturing processes using electron beams are as follows:

- 1) the stopping power of energetic particles, here electrons, in solid (granular) matter, resulting in the interaction energy efficiency;
- 2) secondary phenomena induced during electron–matter interaction (secondary electron emission, charge flow, etc.) leading to energy absorption;
- 3) the behavior of the fusion zone in E-PBF and the defects induced after solidification.

As they pass through matter, electrons lose their energy in the form of inelastic collisions. The *stopping power* corresponds to the average energy lost by the energetic particle per distance traveled. It depends on the energy of the primary electron and the nature of the material. Bethe (1932) developed a theory to determine the stopping power of electrons, including a simplified formulation for low energies (<1 MeV¹), significant for additive manufacturing:

$$-\frac{dE}{ds} = \left(\frac{1}{4\pi\epsilon_0}\right)^2 q^4 2\pi N_A \frac{\rho Z}{AE_s} \ln\left(\sqrt{\frac{eE}{2I}}\right) = 785 \frac{\rho Z}{AE_s} \ln\left(\frac{1.16E}{I}\right) \quad [3.15]$$

where E represents the energy of the incident beam (eV), s the path covered in the material (Å), q the elementary charge (1.602×10^{-19} C), N_A Avogadro's number (6.022×10^{23} mol⁻¹), ρ the density (g·cm⁻³), Z and A the atomic number and the atomic mass number of the target material, respectively, e the base of the natural logarithm, E_s the instantaneous kinetic energy of the electron at the position s and I the average excitation potential (ionization) of the material. E_s and I are expressed in (eV).

Using Bethe's theory, the ESTAR² database and experimental results, we can represent, for example, the stopping power for tungsten (Figure 3.17). At high energies (approaching 10 MeV), electrons lose some of the energy through braking (*Bremsstrahlung*³) radiation. At very high energies (>100 MeV), the material again becomes very absorbent. For lower energies (<1 MeV), the stopping power is mainly due to the interactions (commonly called “collisions”) between the electrons and atoms making up the target material which result in the heating of the crystal lattice, ultimately inducing the desired phase transformations.

The stopping power gives the energy deposited per unit depth (z , Figure 3.18) in the material until almost all of the energy carried by the beam has been absorbed (Kanaya and Okayama 1972). By taking into account the transverse (circular) section of the beam, we obtain the energy transmitted to the surface per unit of volume, that is, the volume energy density (eV·m⁻³ or J·m⁻³)⁴ described in section 3.1. For additive manufacturing, electron beam (EB) energies are generally close to 60 keV. This compromise makes it possible to slightly increase the stopping

1 1 eV represents the energy gained by an electron accelerated under a potential difference of 1 V.

2 Available at: <https://physics.nist.gov/PhysRefData/Star/Text/ESTAR.html>.

3 *Bremsstrahlung* (in German) represents the radiation emitted during the deflection of the path of a relativistic electron beam, a phenomenon used in synchrotron sources.

4 1 eV = 1.60218×10^{-19} J.

power while reducing the production of X-rays (as secondary phenomenon induced) by *Bremsstrahlung*.

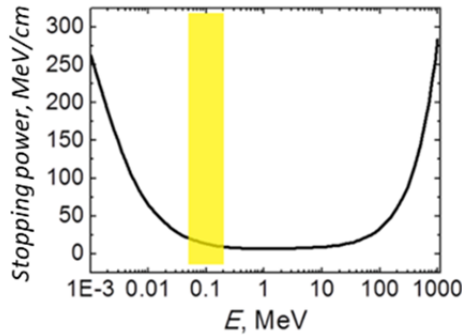


Figure 3.17. Stopping power of W as a function of energy of the primary electron. The colored band indicates the typical EBM energy zone. For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

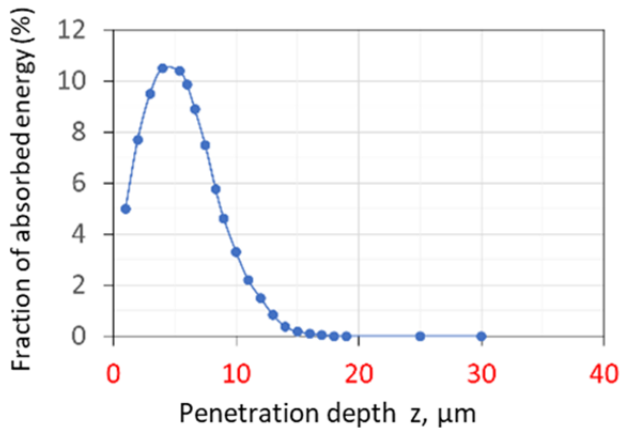


Figure 3.18. Typical fraction of the energy deposited by an EB of 60 keV depending on the depth

3.2.3. The phenomena induced during the electron–matter interaction

When energetic electrons interact with matter, most of the energy is absorbed by the material, whether in solid or liquid phase. It can be deduced from Figure 3.18 that

for powders with a diameter greater than 20 μm , practically all of the energy transported (60 keV) by the EB remains in the irradiated grain (Markl et al. 2013). The other phenomena generated during this interaction are introduced below. Figure 3.19 schematizes the electron–matter interaction. First, many electrons leave the irradiated material. By accumulating all the electrons released during the bombardment of a sample by a beam of energetic electrons according to their kinetic energy, we obtain a characteristic spectrum of the interaction. If the energy of the electrons released from the metal is normalized by that of the incident beam (E_0), we can compare the emission of several materials and several incident beams of different energies.

A characteristic spectrum is shown in Figure 3.20, with three characteristic domains.

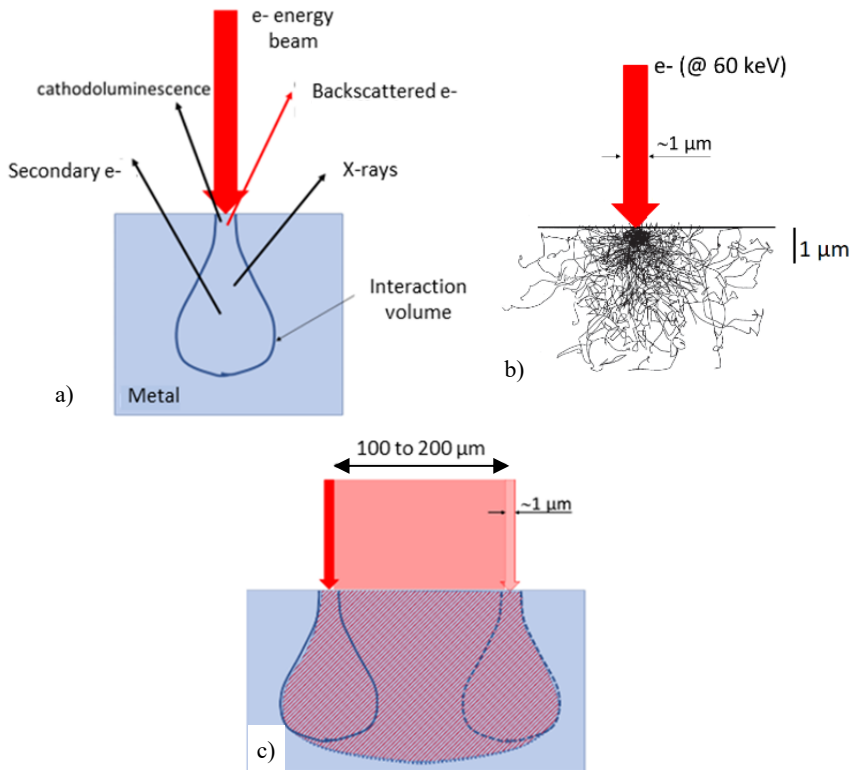


Figure 3.19. a) Phenomena induced during the interaction of an energy EB with a surface; b) simulation of the trajectories of the 60 keV electrons stopped in a solid; c) interaction volume for a typical EB for additive manufacturing. For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

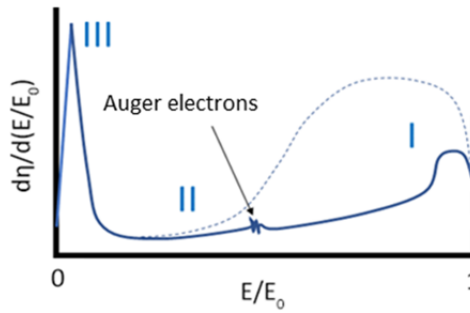


Figure 3.20. Energy spectrum of electrons emitted by a surface irradiated by an EB with energy E_0 . The dotted curve corresponds to a heavy metal; adapted according to Goldstein et al. (2003)

1) Like photons in the case of lasers, some electrons are reflected when they encounter a solid surface. They retain a large fraction of the initial energy (E_0) and are backscattered. For energies between 4 and 40 keV (typical of scanning electron microscopy, Figure 3.19(b)), Hunger and K  chler (1979) gave an analytical expression for the fraction of backscattered electrons:

$$\eta(Z, E_0) = C E_0^m \quad [3.16]$$

where the exponent m is a function inversely proportional to the square root of the atomic number Z and $C < 1$, a coefficient which depends weakly on Z .

In the case of titanium, for example, $Z = 22$, $m = -0.058$ and $C = 0.29$, which gives an incident beam of energy $E_0 = 60$ keV to a fraction of backscattered electrons $\eta \sim 15\%$. Note that the lighter the irradiated element, the lower this coefficient ($\sim 3\%$; zone I, solid curve, Figure 3.20) and conversely, the heavier the element, the more electrons are reflected ($\sim 60\%$; dotted curve, Figure 3.20).

2) The electrons emitted in this interaction can also come from the atoms making up the target material. They are commonly called *secondary electrons*, because they are induced (released) by the primary beam⁵. The greater the incident energy, the closer electrons are to the surface, and the more easily they can exit the metal. Since they must sometimes pass through a substantial thickness (up to μm), the secondary electrons are lower in energy. These electrons come mainly from the external

⁵ Recall that at the quantum level, electrons are indistinguishable and that it is impossible to know whether the electrons leaving the material were initially in the primary beam or in the irradiated material.

orbitals of the atoms in the solid. Some of them exit the material at specific energies (corresponding to their electronic shell structure). The continuous background which connects zone I (high energy) to the low energy part is a mixture between secondary electrons, coming from the solid, and backscattered electrons, which have lost a significant fraction of their energy by crossing a certain material thickness before exiting.

3) The very low energy part of the spectrum (typically <50 eV: zone III, Figure 3.20) is composed only of secondary electrons (Goldstein et al. 2003). These electrons are defined only in relation to their kinetic energy. The secondary emission coefficient is defined as the ratio between the number of electrons released by the surface (n_{SE}) and the number of incident electrons in the beam (n_0). As each electron carries a single elementary electric charge, this coefficient can be written as the ratio of the electron currents emitted (I_{SE}) to the primary current (I_{FE}):

$$\delta(E_0) = \frac{n_{SE}}{n_0} = \frac{I_{SE}}{I_{FE}} \quad [3.17]$$

The secondary emission coefficient is generally less than 1, $\delta < 1$, and less than 0.1 for energies < 50 keV (Reimer and Tollkamp 1980). If the backscattering coefficient is also less than 1, $\eta < 1$, then it becomes clear that most of the energy is deposited in the material. For an electrically insulated surface, even if it is a conductive material, this surface becomes negatively charged under the effect of the incident beam. However, the sum of the electrons emitted, $\eta + \delta$, can, under certain conditions, exceed 1 (especially for energies of a few keV). In this case, the surface becomes positively charged, although it is supplied with electrons. Charge problems arise particularly when the material is insulating (oxidized metal powders, etc.).

The *flow of electric charge* occurs without issue in solid metal, because metals are electrical conductors. However, metal in the granular state (powder) has very different electrical conduction properties from the raw material from which the powder is derived. As an example, consider the case of copper, an excellent electrical conductor in its raw state, characterized by a conductivity of $\sigma_0 \cong 58 \text{ MS} \cdot \text{m}^{-1}$ and density $\rho = 8.96 \text{ g} \cdot \text{cm}^{-3}$. As a large diameter powder (50–100 μm), the bulk density of the powder is only $\rho_{\text{powder}} = 4.6 \text{ g} \cdot \text{cm}^{-3}$, so about half that of solid copper (Raab et al. 2016). In fact, the contact area between the powder grains is very small and separated, and it is almost impossible to ensure continuous contact (percolation). The electrical conductivity of the copper powder then drops to values comparable to electrical insulators, with the oxidation of the powders reinforcing its insulating character. In contrast, for a nickel powder,

which is much less oxidizable, there is a decrease in electrical conductivity of <10 between the solid and the powder.

The manifestation of the electrical insulating character of a metal powder is the accumulation of charges on the powder grains in contact but electrically isolated from the rest of the powder bed (Figure 3.21). If the amount of charge supplied by the primary EB is greater than the pressure drop (flow through a strong contact resistance, secondary emission, etc.), then static electricity is stored on the powder. The charged grains, but without electrical contact, are then subject to a Coulomb repulsive force. Beyond a certain accumulated charge, this repulsion can exceed the force of gravity. In this case, a sudden uplifting of the grains can be seen. This phenomenon is better known as the “smoke” phenomenon, and was modeled by Eschey et al. (2009) (also see Figure 1.33).

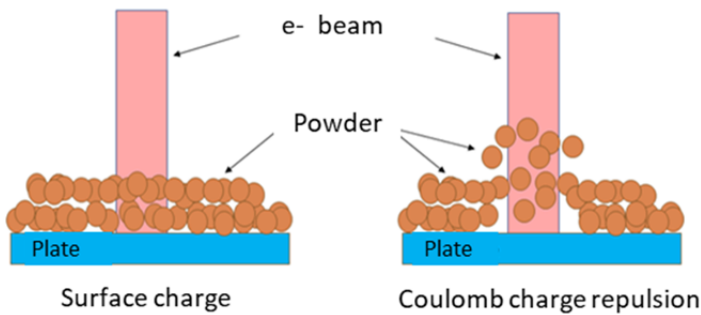


Figure 3.21. *Lifting effect in grains of powder following the accumulation of charge, followed by Coulomb repulsion*

“Smoke” phenomena play a harmful role in the E-PBF process for two reasons: (i) as with laser processes, the presence of powder grains in suspension in the path of the beam (here an electron beam) decreases the energy transmitted to the powder bed and can generate numerous metal vapors (as explained in section 3.2.4); (ii) as the column generating the electron beam operates under vacuum, any rise in pressure in the barrel is detrimental to the filament or to the acceleration and therefore to the shaping of the EBM beam. This can lead to an abrupt shutdown of the electron gun.

3.2.4. Energy absorption in the powder in E-PBF

During the irradiation of a powder grain, considered to be spherical, by an intense beam (typically 60 kV and 10–100 mA) of electrons, specific phenomena

take place. To simplify, consider an EB with the same radius as that of the grain ($r_G \sim 100 \mu\text{m}$) centered on the powder grain. Figure 3.22 shows this interaction and the production of secondary electrons. The surface exposed to primary electrons is close to that resulting from laser irradiation but, in the case of electrons, the penetrated distance is about $20 \mu\text{m}$ (Figures 3.18 and 3.22), whereas, in the case of a laser, it is limited to the skin thickness ($<50 \text{ nm}$ for solid metals). Consequently, the electrons heat the powder grains with the size of the electron beam more in volume and therefore equivalent to the mass. There are two categories of electrons in the primary beam: (i) the electrons captured by the powder grain which give up all their energy (in gray in Figure 3.22) and (ii) the electrons which cross the grain at its periphery (where the bead has a length less than the penetration distance). The paths of these electrons (in red in Figure 3.22) are shown in dotted lines. These electrons can then heat the grains of the layer below by direct irradiation (but at lower energy) and generate more secondary electrons near each surface crossed (Chow et al. 1994). These (dotted blue line) in turn heat the grains nearby.

The electron absorption volume is greater with a laser beam than with an EB for the same beam diameter. Thus, considering a grain with diameter of $r_G = 50 \mu\text{m}$ (Figure 3.22(a)) and a laser energy deposition thickness z of about 50 nm (skin thickness) for a semi-circle, we obtain an interaction volume of $V_{\text{laser}} \approx 2 \pi z (r_G)^2 \approx 39 \mu\text{m}^3$. On the other hand, for an electron beam (Figure 3.22(b)) and an average electron penetration of $15 \mu\text{m}$, the volume over which the energy is distributed in the same grain of powder is $V_{\text{FE}} \approx 59 \times 10^3 \mu\text{m}^3$, about 1,500 times larger than in the case of the laser. This practically explains the absence of a keyhole regime for the E-PBF process.

The interest of E-PBF is evident for powders with a high reflection coefficient (Cu, Al, etc.), which poorly absorb IR laser radiation. Another difference is that during fusion by E-PBF, unlike near infrared laser photons, the absorption of energy is not favored by the transition to the liquid state (Siegel 1976) (e.g. the absorptivity of aluminum goes from 10% to 30% during laser melting). This increase in laser-specific absorption requires the switch to the keyhole regime. Regardless of the grain size, a fraction of the energy is lost as X-ray radiation and via secondary electrons directed to the free surface of the grain. Under optimal conditions, these losses are less than 10%. For comparison, IR laser radiation on a powder bed leads to absorptions of between 37% and 80% depending on the melting regimes (Figure 3.5).

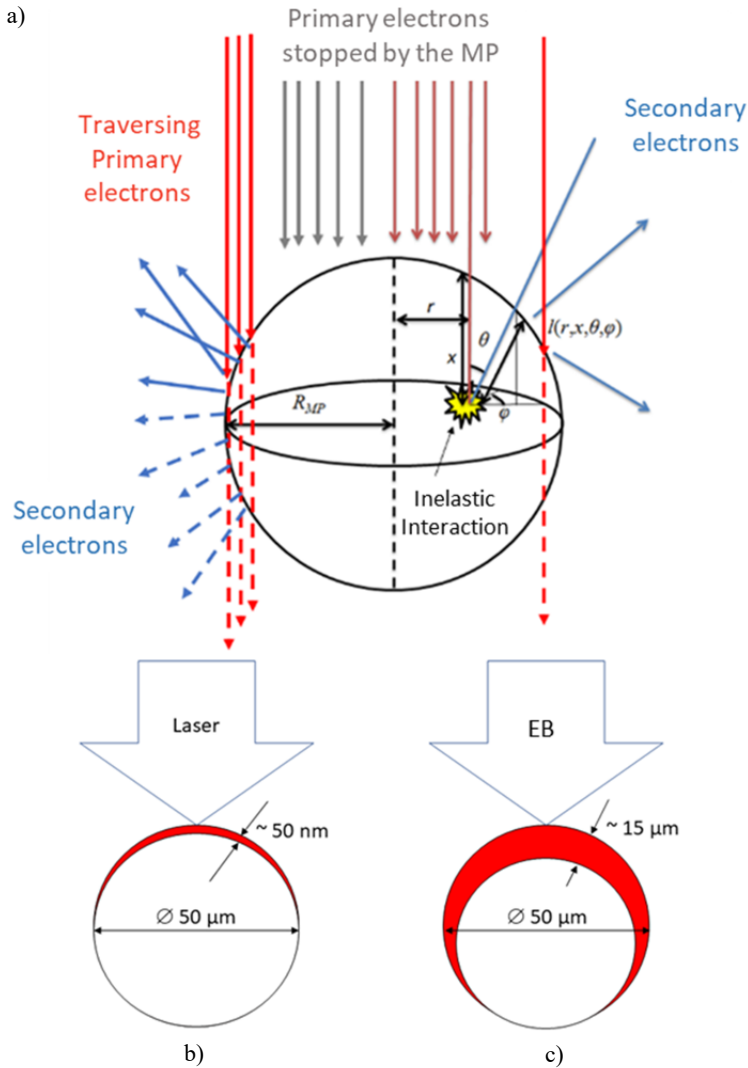


Figure 3.22. a) Production of secondary electrons during irradiation of a microparticle (MP) with a primary EB of the same diameter. (b, c) Comparison of the energy absorption zone (in red) on a grain of $50 \mu\text{m}$ diameter exposed to b) a laser beam and c) an EB of the same diameter. For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

On a powder bed, the charge flows through the weak areas of contact between the grains, characterized by very high electrical resistance (compared to that of the grain volume). The Joule effect then induces localized heating, which leads to microwelds or sintering of the powder grains, unlike L-PBF. The flow from the charge to the mass is generally favored by a preheating step for the powder bed, for example by sweeping the surface rapidly with a very high EB current (see section 3.2.3).

The rise in temperature due to energy transmitted to a surface by an electron beam has been estimated by Castaing (1951) using the formula:

$$\Delta T (K) \sim \frac{4.8 I_{EB} E_0}{K D_0} \quad [3.18]$$

where I_{EB} , E_0 and D_0 , respectively, represent the intensity (μA), the energy (keV) and the diameter (μm) of the incident EB and K the thermal conductivity of the material ($W \cdot cm^{-1} \cdot K^{-1}$).

The fraction of energy deposited on the grain by the electrons (stopping power; see section 3.2.2) leads to heating of the grain until it reaches its melting temperature, or even until the grain evaporates (Seznec et al. 2017).

3.2.5. Description of the fusion zone in E-PBF and associated defects

The powder heating time strongly depends on the interaction time ($t_{int} = D_0/v_0$, quantities defined below) with respect to the conduction time by heat diffusion ($t_{diff} = (l_1)^2/a_{th}$) to the previous layer. If $t_{int} \ll t_{diff}$, then the dissipation of energy is limited by diffusion and the surface becomes overheated, leading to significant evaporation. Conversely, if $t_{int} \gg t_{diff}$, energy is lost by diffusion and fusion is difficult to achieve. For an efficient process, the two times must be of the same order of magnitude, that is, $t_{int} \approx t_{diff}$. In this case, the penetration is optimal, ensuring the link with the previous layer of the homogeneous melt-pool (Figure 3.23). The scan speed (v_0) of the EB is then estimated using:

$$v_0 = \frac{D_0 a_{th}}{\Delta h^2} \quad [3.19]$$

where D_0 is the beam diameter (m), a_{th} is the thermal diffusivity (m^2/s) and Δh is the thickness of the layer to be crossed (m) (Körner et al. 2013).

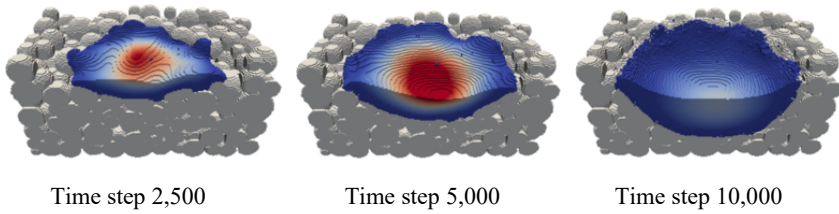


Figure 3.23. Modeling of zone formation from melting by EBM for a 7,500 time-step beam interaction according to Markl et al. (2013). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

However, during the melting of powders, other phenomena come into play. Capillarity is characterized by the Laplace number (La), which represents the cumulative effects of inertia (ρ) and surface tension (γ) with respect to dynamic viscosity (μ). Surface tension is characterized by the Bond number (Bo), which translates the effects of gravity (g) in relation to the effects of surface tension (γ):

$$La = \frac{\gamma l_c \rho}{\mu^2} \quad [3.20]$$

$$Bo = \frac{\rho g l_c^2}{\gamma} \quad [3.21]$$

where γ , μ and ρ are the surface tension, dynamic viscosity and density of the liquid, respectively, l_c is the characteristic length (FZ size) and g is the gravitational acceleration.

For liquid metals such as titanium, the Laplace number is close to $La \sim 10^5$ whereas Bo is much less than 1 ($Bo \sim 10^{-4}$), which means that the surface phenomena are dominant over the viscosity or the gravity. Thus, the liquid metal can effectively wet some adjacent powder particles, but without necessarily melting them, especially if the particles are large. Coalescence between two grains can be limited by inertia or viscosity (Spiegelberg and McKinley 1996). For a characteristic distance $l_c \sim 100 \mu\text{m}$ (FZ size), the final shape of the FZ is rather due to the inertia than to the viscosity of the liquid metal.

The hydrodynamic phenomena remain the same in E-PBF and L-PBF because they are associated with the behavior of the same viscoelastic liquid metal. We present below a comparison of beads obtained by laser fusion and E-PBF

fusion. Figure 3.24 shows the temperature profiles calculated for the E-PBF melting of titanium in a longitudinal direction (that of the movement of the energy source, negative abscissa) and transverse direction (positive abscissa).

The modeling results (Ladani et al. 2017) show that the laser source leads to very localized overheating in front of the FZ (blue dotted curve) while the longitudinal temperature profile presents reduced gradients in the EB (red curve). Likewise, in the transverse direction, the FZ has a more uniform profile with a temperature of 100°C above the melting temperature. It is therefore clear that the EB decreases the heating of the melt-pool compared to the laser beam. Also, the possibility of the keyhole, visible on the left in Figure 3.24, is lower in E-PBF than in L-PBF, essentially because of the volume absorption of energy in E-PBF and the lower level of energy absorbed. However, by adjusting the process parameters (energy and speed), for example, for aluminum (Romano 2015), very similar FZ geometries can be obtained in laser fusion and EB fusion. On the other hand, the injected laser energy must be greater due to the very high reflectivity of aluminum.

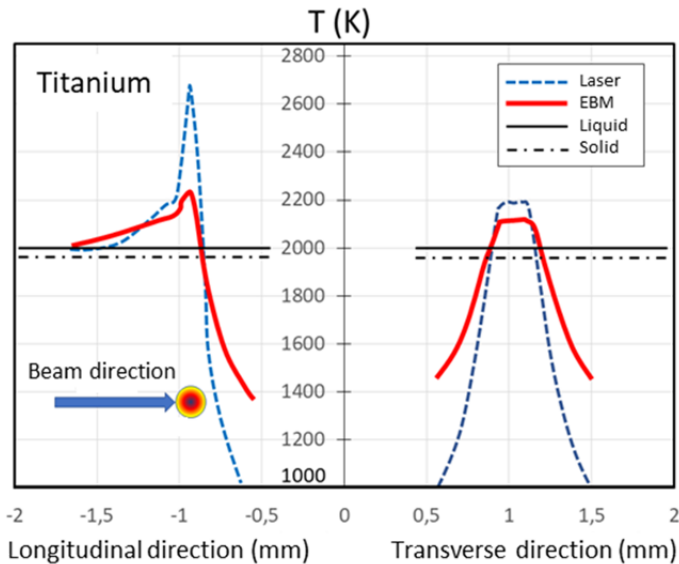


Figure 3.24. Temperature profiles induced in the FZ by laser (blue dotted line) or EB (red line) in the direction of movement of the beam (negative abscissa) and in the transverse direction (positive abscissa) according to Ladani et al. (2017). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

The defects of parts made in E-PBF are generally the same as in L-PBF (see section 3.1.5.2), because they are the consequence of the instabilities of the melt-pool (balling or humping) and not of the way in which the energy is transmitted to the powder bed. We thus find surface defects (humping and balling) that limit the homogeneity of the inter-bead overlaps. Another mechanism leading to the appearance of pores can be the massive evaporation of material, in the presence of strong thermal gradients, especially if the energy deposition is not uniform on the surface of the FZ. In general, the pores remaining in the solid material have dimensions around a few micrometers, or even more if the energy of the beam is not sufficient. The equivalent volume of the porosities is less than 2% for optimal process parameters. From a functional point of view, the presence of these pores does not affect the macroscopic properties of the parts (vacuum resistance, electrical or thermal conductivity, etc.), with the exception of the fatigue properties (see Chapter 3 of Volume 2).

3.2.6. *Partial conclusion regarding E-PBF*

The differences between L-PBF and E-PBF relate mainly to the energy–matter coupling mode and the limitation of the keyhole phenomenon in E-PBF due to there being more volume absorption and lower absorbed energies. In addition, the scanning speed used for E-PBF fusion is generally two to three times greater than with a laser, because of the electromagnetic deflection (without a moving part) of the EB with regard to the mechanical control of the laser optics. Finally, the use of a vacuum in E-PBF can be seen both as an advantage (limitation of oxygen pollution in reactive materials) and as a disadvantage (increased time to cool the work chamber, difficulty of vacuum instrumentation).

3.3. Physics of the wire arc additive manufacturing process

The wire arc additive manufacturing (WAAM) process, unlike the processes using a laser or an electron beam, allows a high material deposition rate. The material melted by the electric arc then forms a large melt-pool that can range from 3 to 20 mm, subject to complex hydrodynamic phenomena. This section recalls the main physical phenomena governing the deposition of molten metal wire by WAAM. It uses experimental and numerical results in order to make the reader aware of the main physical phenomena involved during deposition in an additive manufacturing configuration. The main equations for studying the

phenomena of mass and energy transfers are shown. The physics involved during successive deposits are discussed from test campaigns on simple configurations (walls) with emphasis on the geometry and the defects observed in the deposits.

3.3.1. *Reminder of the essential physical variables*

The physical phenomena in WAAM are quite similar to those encountered during successive deposits of metallic materials with a multipass welding process. The mass and energy transfer phenomena of liquid phases are strongly influenced by the geometric configuration and by the control of the arc voltage $U(t)$, the intensity $I(t)$ and the wire velocity V_f controlled by the welding generator. The overlapping of beads in additive manufacturing amplifies the phenomena related to wetting and energy accumulation during the manufacture of massive shapes or shells. The shielding gas has a great influence on the energy levels developed by the process (the voltage required for ionization, for example), on the protection of the liquid phases and on the behavior of the liquid/gas interfaces. Energy transfers directly affect the solidification conditions of the material and then control the grain structures (and microstructures) as well as residual stresses and distortions. The energies brought into play during the process are governed by the conservation equations (recalled in section 3.3.3.1), which can be complex to solve. The energy developed by the process is influenced by many adjustable and controllable process parameters on the welding generators, which affect the waveform of the electrical signals. The deposit is then both defined by the amount of wire unwound and by the values of these parameters, which influence the energy levels. These parameters are preset by the welding stations. Up to 50 parameters can be changed, with some specific to the arc initiation and extinction phase.

Figure 3.25 illustrates all the physical phenomena at the arc and melt-pool level. During the WAAM process, an arc plasma is created between an electrode and the substrate from an ionized shielding gas. The heating produced by the electric arc induces melting of the feed wire. Droplets form and break off and pass through the arc to feed a melt-pool, creating the deposit after solidification during the movement of the torch. The shape and size of the droplets and the melt-pool are closely related to the heat and mass transfers, which depend on the flows in the fluid phases driven by the dynamic equilibria between the force of gravity, surface tension, the Lorentz force and shear induced by the plasma.

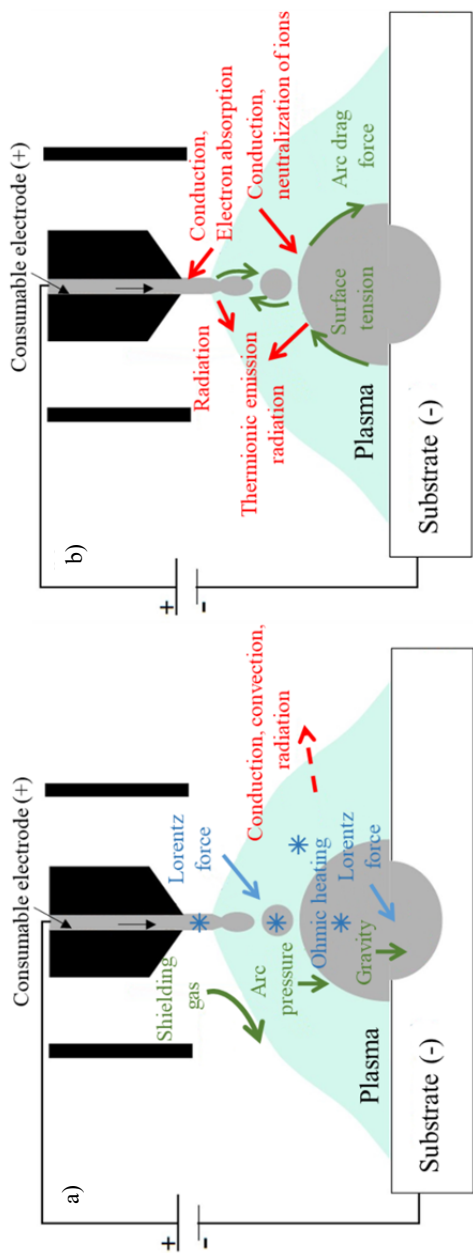


Figure 3.25. Physical phenomena in WAAM : (a) in the solid, liquid and gas phases and (b) at the anode/plasma and cathode/plasma interfaces (Cadiou 2019). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

3.3.2. Arc–wire–deposit interaction

In this section, the physical phenomena induced during the transfer of mass and energy through the arc and the wire are reviewed. Figure 3.25(b) shows the exchanges along different interfaces between the arc and wire and the substrate. Numerical models make it possible to quantify the energies involved more precisely, while describing the dynamic behavior of fluid media during the three phases:

- the formation of the droplet and its detachment at the electrode tip;
- the transfer of the droplet through the arc column;
- the impact of the droplet on the weld pool and its hydrodynamic movements.

3.3.2.1. Physical phenomena in the electric arc

The electric arc is a conductive gaseous medium composed of moving charged particles. Electrons move from the negative terminal (cathode) to the positive terminal (anode); positive ions move in the opposite direction while negative ions move toward the anode. Electron emission occurs at the cathode, which increases with increasing intensity. The anode zone undergoes a bombardment of electrons, which transmit their kinetic energy to the anode in the form of condensation heat, responsible for the strong increase in temperature, which can reach between 5,000 and 20,000 K in the vicinity of the anode, thereby inducing its melting. Most often, direct current is used. In the case of a tungsten inert gas (TIG) process, the tungsten electrode forms the cathode, while the substrate constitutes the anode, which will therefore undergo strong heating, producing a large fusion zone. In the case of an metal inert gas (MIG)/metal active gas (MAG) process, a reverse polarity is used, thus promoting the melting of the electrode. Since most of the energy in MIG/MAG is used to melt the filler metal, the drops are deposited on a slightly preheated substrate, favoring the lack of bonding. In addition, the geometry of the electrode is constantly changing due to melting, which generates a less stable arc.

The energy absorbed by matter can be estimated from the voltage $U(t)$, the current $I(t)$ and the efficiency η , which depend on the different heat losses in the arc and in the molten metal.

3.3.2.2. Arc–wire interaction

With the basic principles of the process detailed in Chapter 1 of this volume, we describe here the physical phenomena involved in the arc and the wire components, and in particular certain couplings such as that between electromagnetism and the behavior of the liquid phases during the detachment of drops. In the MIG/MAG process, but also in TIG, the simplest current modulation corresponds to a pulsed

current with peaks of intensity during a “hot” period and another with a lower intensity (“cold” period). During the hot period, strong currents favor the electromagnetic forces, which act on the detachment of the liquid drops from the wire (Allum 1985). For this modulation, five parameters are already involved (the two periods, the two intensity levels and the wire speed). A large number of WAAM applications use the cold metal transfer (CMT) process. Figure 3.26 shows a synchronized acquisition of the voltage, current and wire speed signals. The control of these signals is done through many process parameters. Chronologically, during the deposition, we notice that the wire plunges with weak electrical signals, then comes into contact with the bead creating a short circuit. The speed slows down and reverses, removing the wire from the melt-pool. The arc starts again, causing the end of the wire to melt.

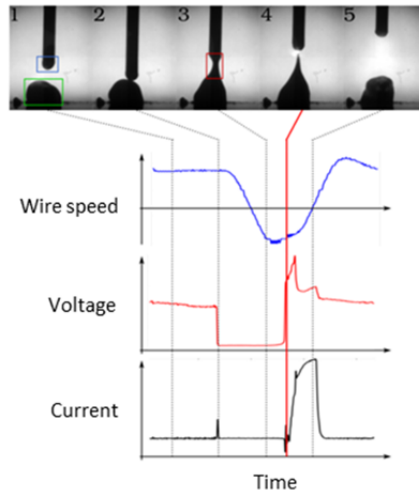


Figure 3.26. Coupled observations of wire speed, voltage and current in the CMT process (red line: short-circuit break; three zones: hanging drop (blue), macro-drop (green) and meniscus (red)). From Monier (2016). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

The physics are similar to those of plasma and are described using the equations presented in Chapter 4 of this volume. A key point is the modeling of the liquid–solid and liquid–gas interfaces. In particular, to describe the formation of drops at the end of the wire, their detachment and their fall into the melt-pool, it is necessary to add a transport equation, as in the case of the level set or volume of fluid methods (detailed in Chapter 4 of this volume).

3.3.2.3. *Detachment mechanisms*

The drop detachment mechanisms are complex and result from a combination of several forces present, evolving over time:

- the force of gravity, which is a force of detachment. It depends on the mass and therefore on the size of the drops, but remains relatively low;
- the drag force induced by the flows within the plasma. It depends on the speed of the shielding gas and tends to accelerate the detached drops;
- the Lorentz electromagnetic force resulting from the passage of electric current and the magnetic field induced by the arc. It constitutes the main force of detachment of the drops, except in the case of CMT;
- the force of surface tension, which opposes the detachment of the drops.

The Lorentz force is significant at high intensities. Its orientation and amplitude depend on the current distribution in the drop (Nemchinsky 1996). This force tends to loosen the drop for divergent current lines or hold it at the end of the wire for converging lines (Figure 3.27). The flow within the liquid drop has no effect on the magnetic field, the Lorentz forces depending only on the shape of the drop and the current distribution (Jones et al. 1998). The behavior of the drop is highly dynamic and an increasing volume increases its inertia. The hanging drop also undergoes oscillations linked to the detachment of the previous drops, which are counterbalanced by the surface tension aimed at making it regain a spherical shape. The oscillations are also damped by the viscous effects.

The size, geometry, detachment frequency and drop rate of the drops depend in particular on operating parameters such as the intensity of the arc, linked to the speed of the feed wire (MIG/MAG). When the current intensity is low (<250 A) and almost constant, the drops are large with a low detachment frequency. When the current exceeds a certain limit, the detachment frequency suddenly increases and the transfer mode changes from globular to axial spray transfer due to Rayleigh–Plateau-type instabilities (Allum 1985). The energy of the arc is then sufficient to quickly melt the end of the wire. As soon as the fine droplets reach a critical size, Lorentz forces produce a pinch-off sufficient to break off the drops. In the globular regime, the large drops formed at the end of the wire break off mainly under the effect of gravity, the Lorentz forces being too weak at the intensities used. The trajectory of the drops is generally random and not necessarily in the axis of the arc, thus generating an unstable transfer with numerous projections.

More stable configurations can be achieved with alternating currents or pulsed currents by imposing current pulses of controlled intensity and duration. The high-intensity phases make it possible to generate a large supply of energy over a short time to promote the melting of the metal and the detachment of the drop, while limiting the heating of the melt-pool thanks to the low intensity phases. In the case of CMT (Figure 3.26), the intensity is first set at a high value for a very short time (a few hundredths of a second), allowing the wire tip to melt and a drop to form. The intensity is then reduced to a value sufficient to maintain the arc. The wire descends and comes into contact with the melt-pool, causing the arc to extinguish. The detachment of the liquid part is then caused by the action of the surface tension and the mechanical rise of the wire and not by the Lorentz forces, which are too weak at low currents. After depositing the droplet in the melt-pool, the arc is restarted by increasing the current, generating a new drop. This process makes it possible to greatly reduce spatter and heat input, because the detachment of the drops is done with a weak current, resulting in low heating of the melt-pool by the Joule effect.

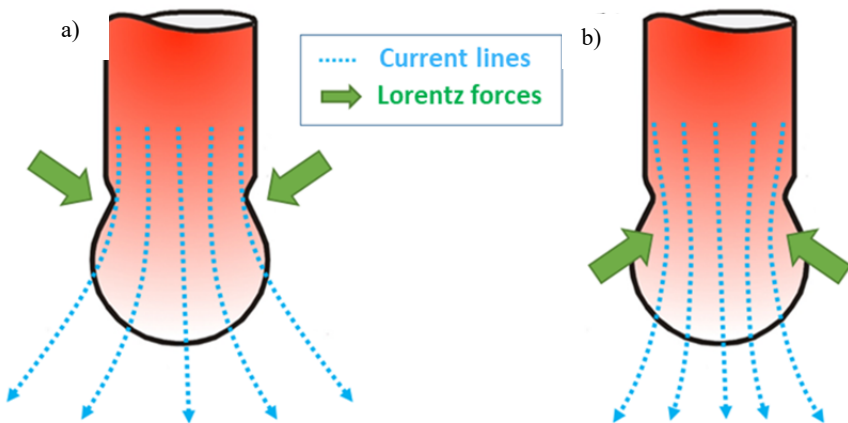


Figure 3.27. Current lines in the drop and in the plasma and Lorentz forces of a) detachment and b) attachment. For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

3.3.2.4. Physical phenomena at the deposit level

Whatever the process, the melting of the deposit can be related to the energy parameters of the process and the amount of mass deposited can be related to the wire speed. For MIG/MAG processes, it is difficult to decouple the input of material

and energy since the electrode is fusible. For these processes, the temporal changes in the electrical signals also result in strong dynamics in the melt-pool. As this melt-pool forms the part during its solidification, the final geometry of the part is therefore strongly impacted by the hydrodynamic movements in the melt-pool. In addition, with the addition of material in the shape of drops and the melting of the substrate or previous layers, the liquid volume changes throughout the manufacturing process. Temperature gradients and velocities within the liquid can be significant. The equations governing the behavior of the deposit are identical to those of the liquid part at the end of the wire from integrating monitoring of moving interfaces.

Figure 3.28 illustrates the temperature and velocity fields during the creation of a deposit, and in particular the effects of the drops as they fall into the weld pool. Different mechanisms control the behavior of liquid phases. The Marangoni effect linked to surface tensions then governs the behavior of certain convection cells. In the case of MIG/MAG, the heating produced by the electric arc induces melting of the feed wire. Drops break off and pass through the arc before reaching the melt-pool. For these liquid parts, the dimensions are much less than 2 mm, which implies the need to take into account magneto-hydrodynamic phenomena (Lorentz forces) to understand the behavior of the liquid.

During the deposition phase, the conservation of the mass makes it possible to link the process parameters (wire diameter d_{wir} , wire velocity V_f , torch travel speed V_s) at the bead surface:

$$S_{\text{bead}} = \frac{\pi d_{\text{wir}}^2 V_f}{4 V_s} \quad [3.22]$$

The shape of the deposit can be calculated using conservation equations, as in Cadiou (2019), or approximated by different geometries (parabola, truncated sphere, etc.). If the width of the molten bath is known as a function of the process parameters, the geometry of the deposit can then be determined. Estimating the bead geometry is necessary to predict the layer height and thus prepare the trajectories. Calculating the bead geometry as a function of the parameters is often difficult and it is often estimated from experimental designs (Almeida 2012), which must be adapted for each material. Some models show interesting predictions (Ríos et al. 2018). These models or experimental designs are often suitable for substrate deposition configurations or for large parts without complex parts and tend to diverge for medium-sized parts.

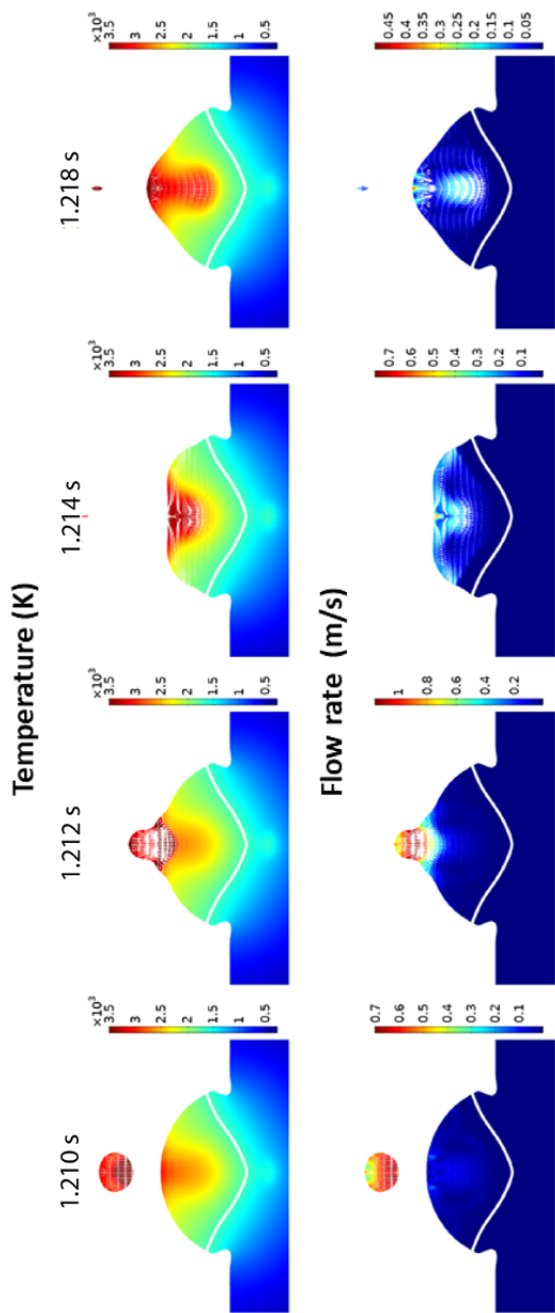


Figure 3.28. Temperature and velocity fields for falling drops in the liquid bath during a pulsed MIG process (Cadiou 2019). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

One advantage of WAAM is that it produces shapes without support because of the predominance of surface tensions over gravity. The Bond number is used to quantify the relationship between the effects related to gravity and those related to surface tension. Formula [3.21], applied to steel, shows that gravity plays an important role for beads with widths of more than 6 mm.

For the deposition phase, carried out at constant torch speed, a synergy is chosen to ensure the good detachment of the liquid phases. The group of settings used is called an operating point. Figure 3.29 shows two deposits made with a CMT (MIG/MAG) station and a feed rate of $900 \text{ mm} \cdot \text{min}^{-1}$ for an aluminum wire of 1.2 mm in diameter, but with different operating points. The difference in geometry comes from a slightly different wire speed, but above all from a longer I_{boost} intensity time at the P7 operating point (Gomez 2018). Applying a higher I_{boost} intensity during a CMT cycle provides more energy and melts a larger surface area of the substrate for better spreading. With less energy, the width of the melt-pool is narrower and the bead more rounded. Despite appearances, the surfaces of the two beads are, apart from measurement errors, within the ratio of the feeding rates.

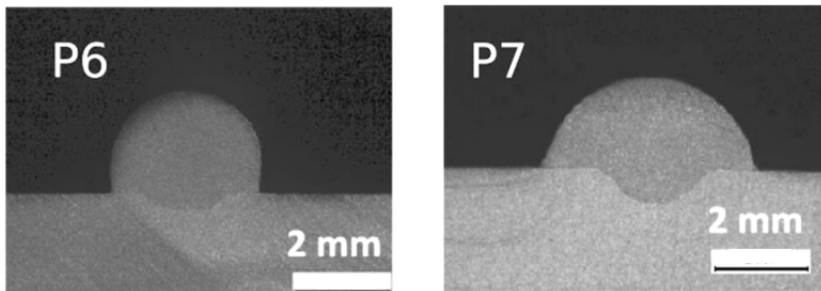


Figure 3.29. Cross-sections of cords obtained with two operating points (Gomez 2018)

During the superposition of the different deposits, depending on the energies involved, the previous layer can be partially remelted. Figure 3.30(a) shows the effect of too much energy on the geometry of the deposit. For the same deposited volume, the higher the temperature of the fusion zone, the more the bead width increases and the height decreases. In addition, the more the surface tension decreases with temperature, the more the liquid phases tend to fragment. Energy transfers are also different if a solid or shell structure is manufactured.

Figure 3.30(b) shows a diagram of thermal diffusion (q corresponding to the thermal flow) in the part being manufactured. Heat removal is lower in the middle figure and the energy supply needs to be better controlled. This type of phenomenon also takes place at the edges of parts for which the heat dissipation is lower. For medium-sized parts, the temperature of the substrate increases significantly as the layers are deposited during construction. Depending on the material, if the same amount of energy is then supplied, the melt-pool will be wider and the height will be lower.

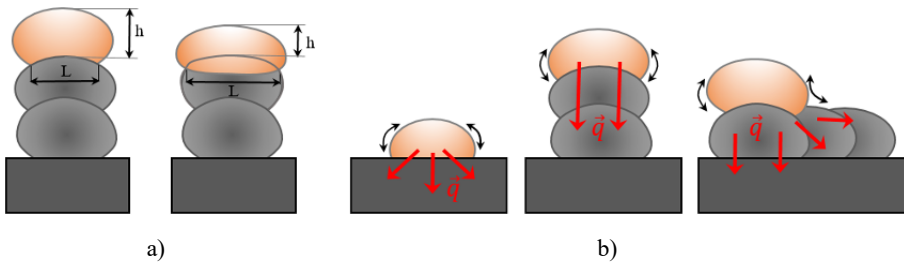


Figure 3.30. a) Effect of energy and wetting on the shape of the bead; on the right, the higher energy induces a larger melt area on the substrate. b) Effect of thermal pumping during manufacture (red arrow: heat flow; black arrow: surface tension). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

In the end, mastering the geometries of the manufactured areas often requires the implementation of control devices.

3.3.3. Form of deposits and associated defects

3.3.3.1. Initiation and extinction of beads

When depositing material, three phases can be distinguished: initiation, deposition and extinction. Process parameters can be divided into two classes. The first include the feed rate and the synergy that depends on the material, gas, wire diameter and wire speed. The second are modifiable parameters for one of the three phases. Apart from the initial and final phases, the deposition takes place with periodic signals during welding (Figure 3.26). Some stations allow the correction of drifts in these phases. For the ignition phase, it is customary in TIG welding to hold the torch in place for a period of time to ensure the melting of the substrate or the previous deposit. This time must be set correctly, otherwise the fusion zone will be too large and will tend to suck liquid, according to the Young–Laplace law in which

a larger radius (R_1 or R_2) of curvature implies lower capillary pressure ($P = \gamma \left(\frac{1}{R_1} + \frac{1}{R_2} \right)$).

On the other hand, in an automated welding process, the arc is stopped when the end position is reached. Figure 3.31(a) shows the anomalies that can be created with too much energy during priming (start of bead) and a lack of material deposited in the final position (end of bead). When the layers are cumulated to make a wall, the geometry shown in Figure 3.31(b) is obtained. It corresponds to an accumulation of the phenomena shown in Figure 3.31(a). For this wall of modest size, the pumping is reduced at the end of the bead, which can lead to a lower height. For more complex parts, the heat diffuses more evenly and this kind of phenomenon is greatly reduced.

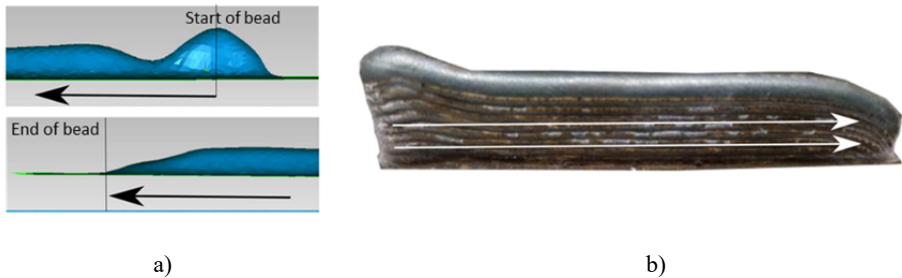


Figure 3.31. a) Non-compliant geometries at the start and end of the bead (Gomez 2018); b) deposit defects for a one-way construction

3.3.3.2. Massive parts

When producing massive parts comprising several juxtaposed passes, the geometry of the bead deposited after solidification is affected both by the shape of the deposition zone and by the physical forces, which act on the material that is still liquid. We generally observe a greater wettability of the melt-pool toward the beads already removed, inducing a decentering of the new deposit with respect to the axis of the filler wire (Figure 3.32). This decentering can lead to the creation of significant hollows between the deposited beads and give rise to filling defects (cavities) or geometrical defects, which are difficult to compensate for by adding new deposits. It is therefore essential to manage the position of the wire in relation to the deposits already formed during construction.

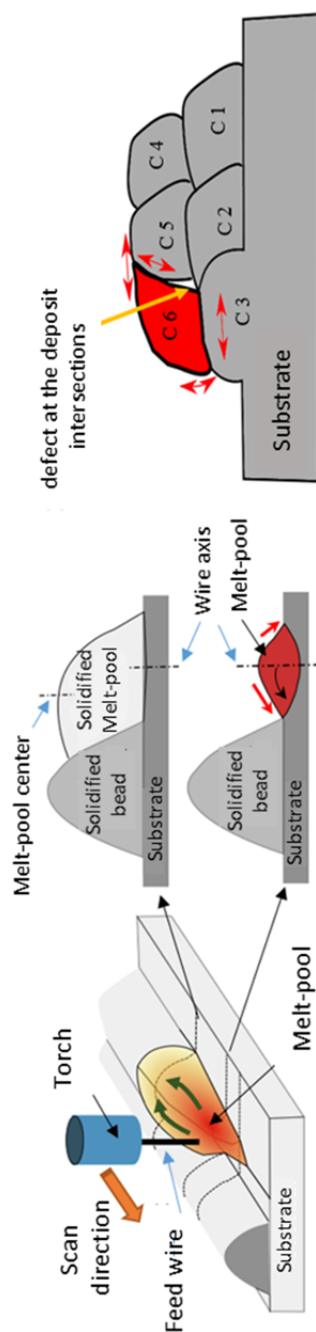


Figure 3.32. Decentering of the bead in relation to the deposit zone linked to the hydrodynamic phenomena that can lead to cavities, according to Li et al. (2018). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

3.3.3.3. Defects related to instabilities

It is well-known in welding that the welding speed is limited by the thermohydrodynamic instability of the melt-pool: humping (Chapuis 2011). This instability linked to the important aspect ratio L_{ZF}/l_{ZF} is detectable by the occurrence of humps and valleys at the level of the bead (Figure 3.33). Although the characteristic lengths are two orders of magnitude greater (10 mm vs. 0.1 mm) in arc welding compared to laser powder bed fusion, the associated physics are similar.

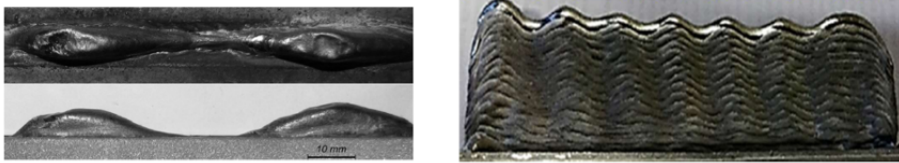


Figure 3.33. Humping phenomena in welding (Chapuis 2011) and during the additive manufacturing of a wall with the MIG CMT process (Gomez 2018)

3.4. Conclusion

Complementary to Chapter 1, this chapter reviewed the different physical phenomena involved in metal additive manufacturing processes using a laser (L-PBF, DED), an electron beam (E-PBF) or an electric arc (WAAM) as a source of energy. It focused on measuring all the complexity of these new techniques for manufacturing 3D objects, which involve multiphysical processes and call on notions of thermal properties, hydrodynamics and energy/matter interaction.

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4

Numerical Simulation of Additive Manufacturing Processes

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The manufacturing strategy has a strong impact on the quality and geometry of parts formed in additive manufacturing. Numerical simulation can be a real help in the choice of manufacturing strategies, minimizing defects and avoiding an often very expensive trial and error process. These tools make it possible to make the link between the operating parameters and the final state of the part, but their development is particularly delicate due to the multiphysics and multiscale aspects present in additive manufacturing. This chapter presents the main equations used to simulate the thermo-hydrodynamic behavior at the scale of the melt-pool as well as the thermo-metallurgical and mechanical behavior at the part scale. Examples will illustrate the capabilities of the models and their interest.

4.1. Thermo-hydrodynamic simulation

4.1.1. *Description of physical phenomena*

Many physical phenomena are involved in additive manufacturing processes (Figure 4.1). The three heat transfer modes are present, with conduction phenomena

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in the solid phase, convection in the fluid phases (liquid metal and gas) and radiation at the surface of the melt-pool. Convection, especially in the melt-pool, requires taking into account the flows that lead to redistribution of energy and therefore modification of the temperature field, melt-pool shapes and material deposition.

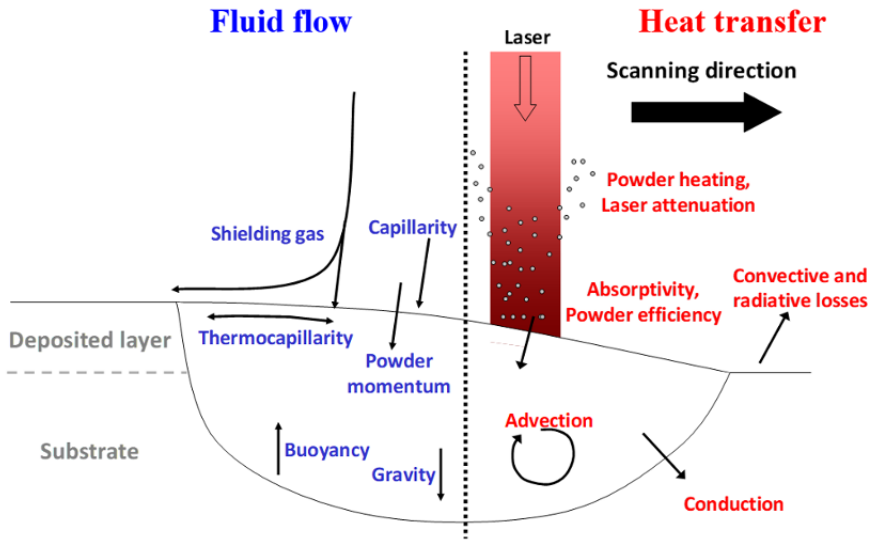


Figure 4.1. Examples of physical phenomena in laser metal deposition (Morville 2012). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

Locally, during the construction of the part layer by layer, the material undergoes complex thermal cycles with periods of heating and cooling, which may give rise to a very heterogeneous microstructure. These thermal cycles also lead to deformations and residual stresses, which may alter the quality of the manufactured part. The modeling and simulation of additive manufacturing processes are therefore particularly complex to implement, since it is necessary to understand the coupling between the different physical phenomena and their scales, which can be very different. In the case of the L-PBF process, a laser beam with a diameter of several hundred microns scans a powder bed at a speed of between 0.1 and $1 \text{ m}\cdot\text{s}^{-1}$ over several square centimeters. The powder particles of several tens of micrometers are distributed over layers of thickness varying between 50 and $100 \text{ }\mu\text{m}$. These multiscale and multiphysic models are therefore very often based on assumptions that are important to make in order to simplify the solving process and not unnecessarily burden the calculations. Simplification aims to reduce the

interactions between physical phenomena and neglect certain phenomena. Thus, in the case of thermo-hydrodynamic modeling, we only deal with the heat transfer and fluid flow phenomena. Metallurgical transformations, strains and stresses are not discussed here. To understand multiscale modeling, it may be interesting to develop models at each scale and to couple these different models in order to reduce calculation times.

4.1.2. Modeling of heat source

To simulate these processes, several approaches can be used. The most physical simulates the phenomenon generating the heat input. For a laser, it involves solving the Maxwell equations associated with this radiation, which makes it possible to take into account the multiple reflections of the laser beam on the wall of the vapor capillary (Courtois et al. 2013). Too restrictive numerically, this method is rarely considered. A second approach represents the process by its energy input. For example, a laser beam or an electric arc is described as a heat source of a thermal problem. In this case, the effects of this heat (flow in the liquid) must be integrated into the simulation by solving the associated equations. For the laser, there is also a third approach, which constitutes a compromise between the two preceding ones. The laser beam is discretized into a finite number of rays each carrying a fraction of the incident power. The rays penetrate inside the vapor capillary (if present), and interact with the melt-pool following the laws of geometric optics. This is the so-called ray-tracing method (Mayi 2019). Less expensive than the first approach, it nevertheless makes it possible to calculate a heat source, taking into account the multiple reflections of the laser inside the capillary. Finally, the fourth and final approach, called the equivalent heat source, is the most simplified. For example, to avoid simulating a fluid flow problem, the heat source can be artificially “stretched” in the manufacturing direction to mimic the effect of the flow on the fusion zone.

4.1.2.1. Process energy input method

The surface heat sources have, in this case, “shapes” based on the spatial distributions of common laser (or electron) beams:

– Gaussian beam:

$$I_0 \left[\frac{W}{m^2} \right] = \frac{2AP_0}{\pi R_{laser}^2} e^{-\frac{2r^2}{R_{laser}^2}} \quad [4.1]$$

– top-hat beam:

$$I_0 \left[\frac{W}{m^2} \right] = \frac{AP_0}{\pi R_{laser}^2} \quad \text{if } r^2 < R_{laser}^2 \quad [4.2]$$

This formulation, classic in laser optics, is stated in $1/e^2$. In this case, the displayed radius of the beam (R_{laser}) is the dimension including intensities greater than 13.5% of the maximum intensity. For the simulation of the LMD process, the masking effect due to the powder jet can be integrated analytically into the previous expressions. To limit the risks of imprecision, optical or thermal tools allow these distributions to be measured experimentally (beam analyzers).

Note that in powder bed additive manufacturing, the presence of powder involves multiple reflections from the laser. If the simulation does not include the detailed description of the powder particles, the previous sources must be adapted to take into account this notion of power density deposition. Different volume absorption models exist; nevertheless, the Beer–Lambert law is a good compromise between simplicity and representativeness. However, two cases should be distinguished: first, when the beam irradiates the powder (at the start of the laser/matter interaction and in front of a beam which advances) and, second, when the beam is on the liquid. In this last case, the source must be described by the preceding surface relations.

4.1.2.2. *Equivalent heat source method*

The use of an equivalent heat source is preferred for thermomechanical simulations, when the fluid simulation becomes too restrictive numerically. In this case, the heat input must be representative of the penetration of the beam into the vapor capillary (if present) and the melt-pool hydrodynamics. There is therefore an infinite number of possible heat sources depending on the operating conditions. Parameterized sources such as the cylindrical involution normal (CIN) (Ranatowski and Ciechacki 2010) or the Goldak source (Goldak et al. 1986) are preferred. The CIN source is formulated according to:

$$S_{CIN} \left[\frac{W}{m^3} \right] = \frac{AP_0 k K_z}{\pi (1 - e^{-K_z e_s})} e^{-k(x^2 + y^2) - K_z z} [1 - H(z - e_s)] \quad [4.3]$$

where k is the concentration factor, K_z is the involution factor, e_s is the application thickness and $H()$ is the Heaviside function. To integrate the flow effects, some consider an anisotropic thermal conductivity, enhanced in the scanning direction, at the rear of the heat source (4–100 times the conductivity of the solid phase). The Goldak heat source model incorporates this effect through its front/rear asymmetry:

– front part:

$$S_f \left[\frac{W}{m^3} \right] = \frac{6\sqrt{3}AP_0}{A_f B_s C_s \pi \sqrt{\pi}} e^{(-3\frac{x^2}{A_f^2} - 3\frac{y^2}{B_s^2} - 3\frac{z^2}{C_s^2})} \quad [4.4]$$

– rear part:

$$S_r \left[\frac{W}{m^3} \right] = \frac{6\sqrt{3}AP_0}{A_r B_s C_s \pi \sqrt{\pi}} e^{(-3\frac{x^2}{A_r^2} - 3\frac{y^2}{B_s^2} - 3\frac{z^2}{C_s^2})} \quad [4.5]$$

where A_f, A_r, B_s and C_s are the parameters of the distributions, respectively, along x in front of the source or at the rear of the source, y and z . Note that the parameters ($k, K_z, e_s, A_f, A_r, B_s$ and C_s) can be imposed *a priori* or identified experimentally. Many identification methods exist, such as the Levenberg–Marquardt method, which is robust, not very sensitive to measurement noise and allows identification by minimizing a quadratic deviation between measured and calculated temperatures.

4.1.2.3. Multiphysics approach: WAAM

In the particular case of the WAAM process, if one wishes to avoid an “equivalent source” approach, which requires the identification of the heat source parameters using experimental data (an expensive procedure), it is possible to re-run the development of complete multiphysics models integrating the entire physics of the arc. It is then necessary to solve in all media (plasma, wire, substrate) the conservation equations allowing the description of all the magneto-hydrodynamic phenomena involved:

$$\rho c_p \left[\frac{\partial T}{\partial t} + \vec{u} \cdot \nabla T \right] = \nabla \cdot (K \nabla T) + \vec{j} \cdot \overrightarrow{E_{elec}} - S_{ray} + S_v \quad [4.6]$$

$$\frac{\partial \rho}{\partial t} + \nabla \cdot (\rho \vec{u}) = 0 \quad [4.7]$$

$$\rho \left(\frac{\partial \vec{u}}{\partial t} + (\nabla \vec{u}) \vec{u} \right) = \nabla \cdot [-p I_d + s] + \rho \vec{g} + \vec{j} \times \vec{B} + \vec{f}_v \quad [4.8]$$

$$\nabla \times \left(\frac{1}{\mu_m} \nabla \times \overrightarrow{A_m} \right) + \sigma_0 \nabla V = \vec{0} \quad [4.9]$$

$$\nabla \cdot (\sigma_0 \nabla K) = 0 \quad [4.10]$$

with $s = 2\mu \left(\dot{\epsilon} - \frac{1}{3} \text{tr}(\dot{\epsilon}) I_d \right)$, $\dot{\epsilon} = \frac{1}{2} (\nabla \vec{u} + (\nabla \vec{u})^T)$ and $\text{tr}(\dot{\epsilon}) = \nabla \cdot \vec{u} = \frac{-1}{\rho} \frac{d\rho}{dt}$, where ρ denotes the density, c_p is the heat capacity, K is the thermal conductivity, T is the temperature, $\vec{j}$ is the current density, $\overrightarrow{E_{elec}}$ is the electric field, S_{ray} is the radiation

losses in the arc, S_v is a source term representing the heat fluxes exchanged at the level of free surfaces, $\vec{u}$ is the velocity, p is the pressure, I_d is the identity matrix, μ is the dynamic viscosity, $\vec{g}$ is the acceleration of gravity, $\vec{B}$ is the magnetic field, $\vec{A}_m$ is the magnetic potential, V is the electric potential, μ_m is the magnetic permeability of the medium and σ_0 is the electrical conductivity.

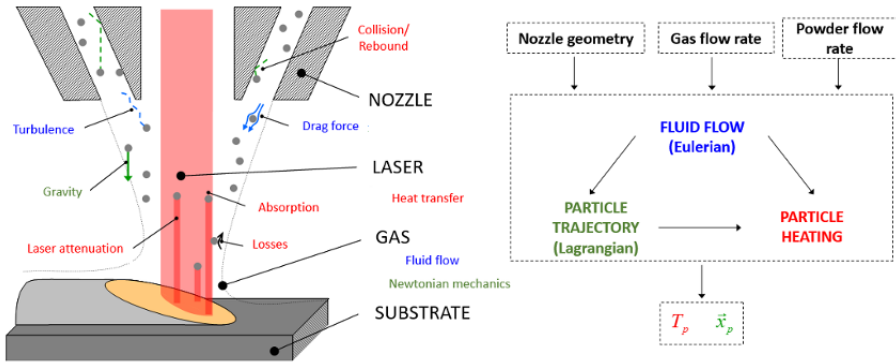


Figure 4.2. Illustration of a powder jet model (Morville 2012). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

Plasma is generally assumed to be in local thermodynamic equilibrium. This assumption is however false in the vicinity of the plasma–electrode interfaces. We then introduce specific heat fluxes reflecting the energies involved in the emission or absorption of electrons and positive ions (Zhao and Chung 2017). These terms are taken into account in equation [4.6] via the term S_v . With this multiphysics approach, the energy from the arc results from both the term $\vec{j} \cdot \vec{E}$ representing the Joule effect and the term S_v . This type of approach makes it possible to have a completely predictive approach, since it does not require any hypothesis on the spatial and temporal distribution of the heat source. More details are available in Cadiou (2019).

4.1.3. Modeling of material input

It is important to precisely determine the amount of material addition in the melt-pool, since it conditions the volume of the deposit and therefore the final geometry of the part. Several approaches are possible depending on the process. If the material arrives in powder form, it is possible to model the contribution with a set of discrete undeformable particles of uniform temperature to predict their trajectory,

their speed and their temperature before impact in the melt-pool (as in the case of the LMD process, see section 4.1.3.1). The assumption of a uniform temperature is justified if the Biot number (the product of the convective exchange coefficient with the characteristic size of the particles divided by the thermal conductivity of the particle) is much lower than 1. For the other cases where significant thermal gradients are present, the filler material should be modeled as a deformable continuous medium, which may be the case for a filler wire. For the L-PBF process, we can highlight models at the powder particle scale, simulating the coalescence of molten particles and continuous models (homogenization model).

4.1.3.1. Supply by powder projection modeled with a discrete approach

The powder jet models make it possible to describe the gas flows at the level of the nozzle, the trajectory of the powder particles before entering the melt-pool as well as their heating with the laser. These models, as illustrated in Figure 4.2, therefore make it possible to determine the spatial distribution of the powder particles and their temperature before entering the melt-pool. These data can then be used as input data for the models of the weld pool.

The trajectory of the supposedly spherical powder particles is calculated using a balance of forces involving the force exerted by the gas and gravity (Morville 2012):

$$m_p \frac{d \vec{v}_p}{dt} = \frac{18 \mu}{\rho_p d_p^2} \frac{C_D Re_p}{24} m_p (\vec{u}_g - \vec{v}_p) + m_p \vec{g} \quad [4.11]$$

where m_p is the mass of the particle, t is the time, $\vec{v}_p$ and $\vec{u}_g$ are the velocity vectors of the particle and the gas respectively, μ is the dynamic viscosity of the gas, C_D is the drag coefficient, Re_p is the Reynolds number of the particle ($Re_p = (\rho_g |\vec{u}_g - \vec{v}_p| d_p) / \mu$), ρ_p is the density of the particle, d_p is the diameter of the particle and $\vec{g}$ is the acceleration of gravity. The temperature of the spherical particles assumed to be uniform is described by an energy balance taking into account the heat input from the laser and the losses from convection and radiation (Morville 2012):

$$m_p c_p \frac{dT_p}{dt} = A I_0 \frac{\pi d_p^2}{4} - [h_c (T_p - T_\infty) + \varepsilon_p \sigma_{SB} (T_p^4 - T_\infty^4)] \pi d_p^2 \quad [4.12]$$

where c_p is heat capacity at constant pressure, T_p is the temperature of the particle, d_p is the diameter, A is the absorptivity, ε_p is the emissivity, I_0 is the surface density of the laser beam incident on the surface of the particle, h_c is the convective heat transfer coefficient, σ_{SB} is the Stefan–Boltzmann constant and T_∞ is the temperature of the surrounding medium.

4.1.3.2. *Supply with melted wire*

In the case of a feed wire, its speed and diameter are known. It is then possible to quantify the flow rate of material supplied to the melt-pool:

$$q_m = \rho_{fil} S_{fil} V_f \quad [4.13]$$

where ρ_{fil} is the density of the wire at room temperature, S_{fil} is the cross-section of the wire and V_f is its velocity. This mass flow in $\text{kg}\cdot\text{s}^{-1}$ defines a source term in the continuity equation, which is described in section 4.1.5. Finer modeling is also possible. It consists of simulating the melting of the filler wire, the formation of drops and their fall into the melt-pool. Specific methods of interface capture are then necessary. Some are presented below.

4.1.4. *Numerical methods for deposition modeling*

In order to simulate the layer-by-layer construction of the part, we distinguish between two types of approaches. One assumes the shape of the deposits is known. The methods associated with this approach is described in section 4.2. The other approach is to predict the shape of deposits based on operating parameters. This approach, which is more expensive in terms of computation time, is particularly interesting for understanding the appearance of defects such as surface finish, porosities and the balling phenomenon present in L-PBF. Among the possible methods, a distinction should be made between fixed-mesh methods and moving-mesh methods. The methods developed for moving boundary problems make it possible to predict the displacement of the free surface as a function of the supply of material and the various forces present, such as surface tension.

4.1.4.1. *Fixed-mesh methods: level set and volume of fluid*

Level set or volume of fluid (VOF)-type methods are particularly interesting, because they make it possible to simulate the melting of the filler wire, the fall of drops into melt-pool, the growth of the deposit, the coalescence of the molten particles and the balling phenomenon. For these methods, the material and the surrounding gas are assimilated to a two-phase medium. A variable is introduced in the whole computational domain to identify the position of the material/gas interface. A transport equation associated with this variable allows the calculation of the displacement of the interface over time as a function of the velocity.

In the case of the level set method, introduced by Osher and Sethian (1988), a distance function at the interface, denoted ϕ , varies continuously from 0 (for example in gas) to 1 (in matter) (Figure 4.3(a)). The 0.5 isovalue marks the position of the material/gas interface. The ϕ function is also used to assign the appropriate properties in each domain. Thus, for the density, we define:

$$\rho = (\rho_{mat} - \rho_{gas}) \phi + \rho_{gas} \quad [4.14]$$

The displacement of the interface is obtained using a transport equation related to the speed of the fluids $\vec{u}$:

$$\frac{\partial \phi}{\partial t} + \vec{u} \cdot \nabla \phi = 0 \quad [4.15]$$

This equation can be modified to reset the function ϕ to ensure the good behavior of the interface over time (Courtois et al. 2016; Chen 2018). The function ϕ makes it possible to define a Dirac function of the form $\delta(\phi) = C_{LS} \phi(1 - \phi) |\nabla \phi|$, where C_{LS} is a constant, non-zero in the vicinity of the interface, allowing all the interfacial conditions to be introduced as source terms in the conservation equations.

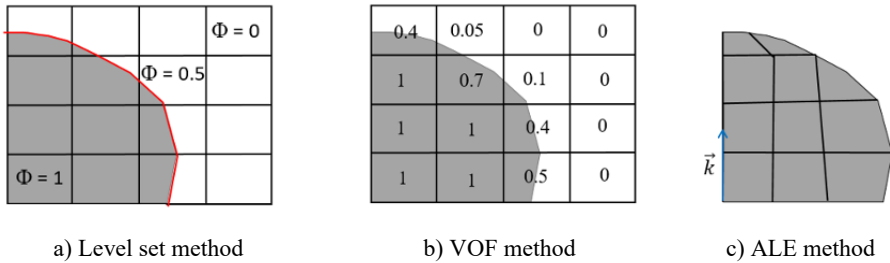


Figure 4.3. Numerical methods used to simulate material deposition in additive manufacturing. For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

The VOF method, proposed by Hirt and Nichols (1981), involves defining, in each cell, a volume fraction F representing the occupancy rate of one of the fluids in the cell (Figure 4.3(b)). This fraction varies between 0 and 1 and satisfies a speed-dependent transport equation. With the level set and VOF methods, it is possible to process the input of matter in two ways: either by explicitly representing each particle using the variables ϕ and F , or by introducing a global source term into the mass conservation equations.

4.1.4.2. Moving mesh methods: arbitrary Lagrangian–Eulerian method

The arbitrary Lagrangian–Eulerian (ALE) method, proposed by Hirt et al. (1974), involves modeling only the substrate and changing its geometry by moving the nodes of the free surface as a function of the amount of material added (Figure 4.3(c)). The nodes inside the domain are moved arbitrarily in order to ensure the good quality of the mesh. In the case of a powder stream process, the speed of the interface $\vec{u}_{LG}$ can be calculated from the speed of the liquid metal on the surface $\vec{u}$ and the velocity linked to the supply of material $\vec{u}_p$:

$$\vec{n} \cdot \vec{u}_{LG} = \vec{n} \cdot \vec{u} + \vec{n} \cdot \vec{u}_p |_{T > T_f} \quad [4.16]$$

The temperature condition on $\vec{u}_p$ makes it possible to take into account a material input only when the surface temperature is higher than the melting temperature T_f . Assuming a Gaussian powder distribution, the velocity $\vec{u}_p$ is given by:

$$\vec{u}_p = N_{jet} \frac{\eta_p D_m}{\rho_p \pi r_{jet}^2} \exp\left(-N_{jet} \frac{r^2}{r_{jet}^2}\right) \vec{e}_z \quad [4.17]$$

where N_{jet} is a constriction coefficient, η_p is the interaction efficiency of the powder jet with the melt-pool, D_m is the mass flow rate of powder, r_{jet} is the radius of the powder jet, r is the radial distance from the axis of the nozzle and $\vec{e}_z$ is a vertical unit vector. These parameters can be deduced from experimental data or from a powder jet model (Morville 2012).

4.1.5. Modeling of heat and mass transfer in the melt-pool

4.1.5.1. Driving forces for fluid flow in the melt-pool

The melt-pool created in additive manufacturing experiences large flows whose driving forces are varied. These different movements strongly condition the energy distribution within the melt-pool and therefore the final shape of the deposit (Figure 4.4). The melt-pool sizes can vary from a few hundred micrometers for the L-PBF process to several millimeters or even centimeters for WAAM, with speeds between 0.1 and 1 m·s⁻¹. Among the driving forces, we can mention the surface hydrodynamic forces, such as the surface tension (equation [4.18]), the Marangoni effect (equation [4.19]), aerodynamic shear and recoil pressure (in the case of vaporization), and the volume forces acting within the melt-pool, such as buoyancy (equation [4.20]) and Lorentz forces (in the case of an electric arc, equation [4.21]).

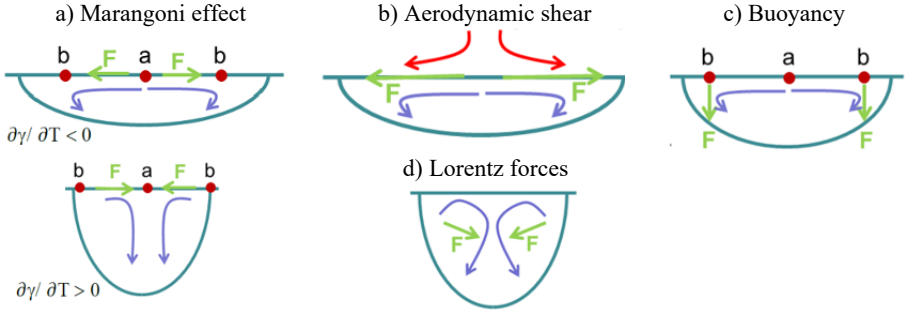


Figure 4.4. Influence of different flow forces on the shape of the melt-pool.
For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

These different forces are governed by the following equations:

$$\vec{f}_\gamma = \gamma \kappa \vec{n} \quad [4.18]$$

$$\vec{f}_m = \frac{\partial \gamma}{\partial T} \nabla_s T \quad [4.19]$$

$$\vec{f}_f = \rho \vec{g} = \rho_f (1 - \beta_v (T - T_f)) \vec{g} \quad [4.20]$$

$$\vec{f}_l = \vec{j} \times \vec{B} \quad [4.21]$$

where γ is the surface tension coefficient, κ is the curvature, $\vec{n}$ is the normal at the interface, $\frac{\partial \gamma}{\partial T}$ is the thermocapillary coefficient, $\nabla_s T$ is the tangential thermal gradient, β_v is the volume expansion coefficient, $\vec{j}$ is the current density vector and $\vec{B}$ is the magnetic field. The buoyancy force has a very limited effect on the temperature field, but it is generally taken into account because of its easy implementation and low computational cost.

4.1.5.2. Conservation equations

In order to simulate the thermo-hydrodynamic phenomena in the melt-pool, it is necessary to solve the energy (equation [4.22]), mass (equation [4.23]) and momentum (equation [4.24]) conservation equations, which, respectively, constitute the heat, continuity and Navier–Stokes equations. They are resolved in the solid, liquid and gas domains:

$$\rho c_p \left[\frac{\partial T}{\partial t} + \vec{u} \cdot \nabla T \right] = \nabla \cdot (K \nabla T) + S_v \quad [4.22]$$

$$\frac{\partial \rho}{\partial t} + \nabla \cdot (\rho \vec{u}) = 0 \quad [4.23]$$

$$\rho \left(\frac{\partial \vec{u}}{\partial t} + (\nabla \vec{u}) \vec{u} \right) = \nabla \cdot [-p I_d + s] + \vec{f}_f + \vec{f}_v \quad [4.24]$$

with $s = 2\mu \left(\dot{\epsilon} - \frac{1}{3} \text{tr}(\dot{\epsilon}) I_d \right)$, $\dot{\epsilon} = \frac{1}{2} (\nabla \vec{u} + (\nabla \vec{u})^T)$ and $\text{tr}(\dot{\epsilon}) = \nabla \cdot \vec{u} = -\frac{1}{\rho} \frac{d\rho}{dt}$, where ρ denotes the density, c_p is the heat capacity, K is the thermal conductivity, S_v is a volume source term (zero if the heat input is integrally applied at the surface), T is the temperature, $\vec{u}$ is the velocity, p is the pressure, I_d is the identity matrix and μ is the dynamic viscosity.

The volume force $\vec{f}_v$ may include the Lorentz forces in the case of a WAAM process, but also the Darcy damping force that allows the velocity field to be canceled out for temperatures below the melting temperature (Morville 2012). It is also possible to use a very high viscosity in the solid phase (Chen 2018). The enthalpy of fusion can be taken into account either using an enthalpy formulation for equation [4.22] (Chen 2018) or using an equivalent specific heat (Morville 2012). In the case of a level set or VOF method, the conditions at the material–gas interface are added as source terms in the conservation equations using the Dirac function, as shown in section 4.1.4.1.

4.1.5.3. Boundary conditions

For the thermal problem, the surface heat input as well as the losses by convection and radiation at the surface of the melt-pool are taken into account:

$$K \nabla T \cdot \vec{n} = \varphi_{source} - h_c(T - T_\infty) - \varepsilon \sigma_{SB}(T^4 - T_\infty^4) \quad [4.25]$$

For other surfaces, radiative and convective losses are considered. Depending on the temperature levels, vaporization losses can be introduced.

For fluid mechanics, the surface tension and Marangoni effect at the surface of the melt-pool are taken into account. Depending on the process, shear forces induced by surrounding gases (shielding gas, plasma) can be introduced using hypotheses or by solving the Navier–Stokes equations for these fluids. The same is true for the arc pressure (WAAM) or recoil pressure (L-PBF) if a vapor capillary is present.

4.1.5.4. Thermophysical properties of materials

Solving the previous equations requires knowledge of the properties of the filler materials and the substrate from room temperature to temperatures close to

evaporating temperature. Due to the measurement difficulties at these temperatures, these data are scarce in the literature, in particular for the liquid phase. Models are often based on assumptions that may limit the reliability of the results. We can nevertheless cite the reference study by Mills (2002), which lists many material properties as a function of temperature.

Devices have also been developed, such as the one proposed by Le Maux (2020), to measure the thermophysical properties of liquid metals. The device illustrated in Figure 4.5 involves aerodynamically levitating a sample in a controlled atmosphere chamber. The sample is heated by a powerful laser to reach temperatures above the melting point. The measurement of temperature and high-temperature emissivity are carried out by multispectral pyrometry. The density measurement is derived by measuring the expansion of the material using a high-speed camera and backlighting. Acoustic excitation in the levitating gas causes the sample to oscillate at its resonant frequency, allowing measurement of surface tension or dynamic viscosity. Figure 4.5 gives an example of surface tension measurements obtained from niobium, for which there are few results in the literature. Such measurements are particularly important to properly model the Marangoni effect, but they can also help in better understanding the instabilities of the melt-pool in relation to the composition of the shielding gas (argon, helium) or the composition of the filler material.

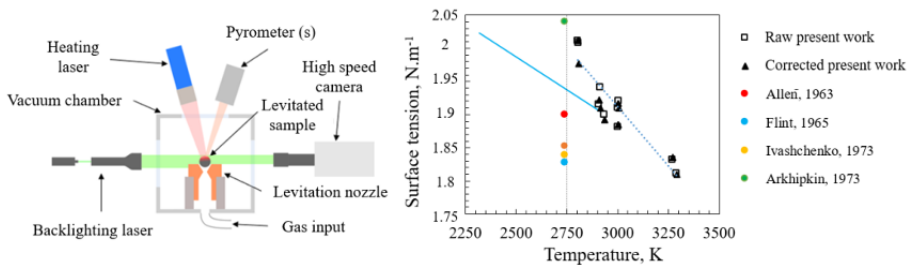


Figure 4.5. Surface tension measurement by aerodynamic levitation. Example of measurements obtained for niobium (Le Maux 2020). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

4.1.6. Examples of thermo-hydrodynamic simulations

In accordance with the operating parameters that condition the kinetics and thermal gradients, the size of the mesh and the time steps are to be chosen with care. Taking

into account the surface tension and the Marangoni effect also introduces strong constraints for the mesh. Furthermore, thermo-hydrodynamic 3D multilayer simulations are still rare because of the computation times. Next, we present simulation examples illustrating the interest of thermo-hydrodynamic simulations.

The first example concerns the build-up of a titanium alloy wall (Ti-6Al-4V) via the LMD process, simulated using the commercial software COMSOL Multiphysics® (Morville 2012). The objective of these simulations was to understand the physical mechanisms responsible for poor surface finish and to propose operating conditions to reduce roughness in order to limit re-machining operations. Two- and three-dimensional models were used to calculate quantities that are difficult to access with experiments, such as the speed and temperature of the particles in the powder jet, the 3D shape of the melt-pool and the speed and temperature within the melt-pool. These self-consistent models have made it possible to predict surface finish from operating parameters (laser power and speed, gas and powder inlet flow rates, nozzle shape, materials). The ALE method was chosen to simulate the formation of different deposits. Figure 4.6 shows the initial geometry of the substrate as well as the speed and temperature fields calculated by a 2D transverse thermo-hydrodynamic model representing a half-wall 1 mm wide (plane of symmetry on the left vertical axis) when building the first layer ($t = 4.5$ s) and the fifth layer ($t = 80.5$ s). The laser source applied to the surface generates a melt-pool subjected to flow velocities of up to $1.4 \text{ m}\cdot\text{s}^{-1}$. Due to a negative thermocapillary coefficient, the Marangoni effect induces maximum fluid velocities at the surface, oriented toward the lateral edge of the wall. Two convection cells are then formed, one of low intensity inside the wall and the other of high intensity, which induces a significant penetration of the melt-pool on the lateral edge. These illustrations clearly show the predominant role of fluid mechanics on the temperature field and consequently the shape of the melt-pool and the deposit. Taking into account the surface tension and the material input makes it possible to predict the undulations on the lateral face of the wall, clearly visible at $t = 80.5$ s. A parameterized study performed with this model showed that the surface finish of a Ti-6Al-4V wall was better at high laser powers, fast feed rates and low mass powder flow rates (Morville 2012).

Figure 4.7 shows velocity and temperature fields obtained using a single-layer 3D model for a 316L stainless steel and two equivalent activity values a_k . As indicated in Morville (2012), the surface tension coefficient depends both on the temperature and on the equivalent activity of the surfactant elements present at the surface of the melt-pool. So, for $a_k = 0.001\%$, we obtain a negative thermocapillary

coefficient, while for $a_k = 0.011\%$, it becomes negative for very high temperatures and positive for low temperatures. The application of the same heat source leads to very different melt-pool and deposit shapes, with the appearance of two convection rollers for the case $a_k = 0.011\%$ and a larger hot area on the surface (Morville 2012).

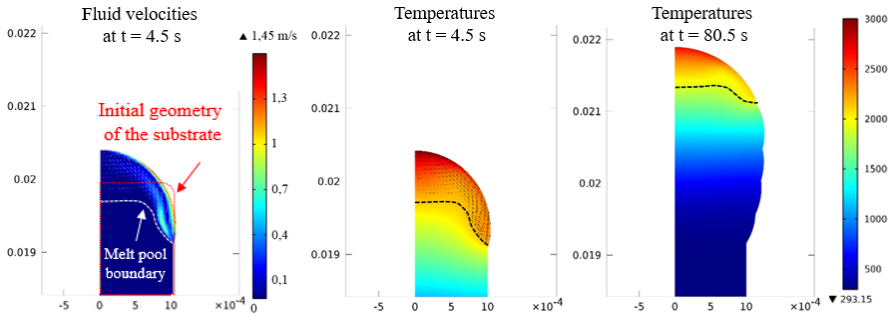


Figure 4.6. Velocity and temperature fields at $t = 4.5$ s (first layer) and $t = 80.5$ s (fifth layer) (Ti-6Al-4V alloy, $P_0 = 320$ W, $V_0 = 0.2$ m·min⁻¹, $D_m = 1$ g·min⁻¹). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

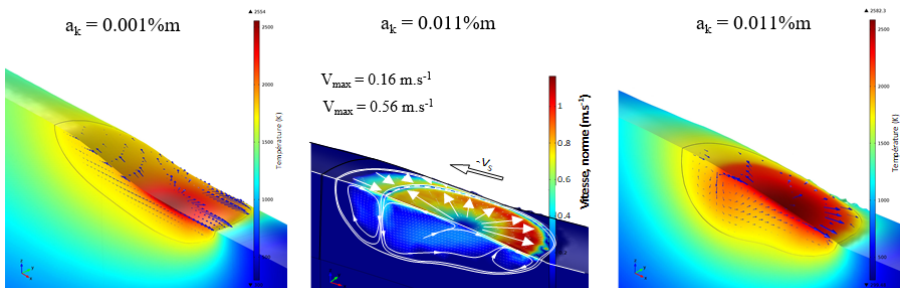


Figure 4.7. Temperature and velocity fields for two activity values a_k (316L stainless steel, $P_0 = 400$ W, $V_0 = 0.4$ m·min⁻¹, $D_m = 2$ g·min⁻¹). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

The second example concerns the simulation of the WAAM process, the equations of which were presented in section 4.1.2.3. Figure 4.8 illustrates the temperature fields from a 3D model simulating a cold metal transfer (CMT) process. The detachment of the drops is then obtained during the mechanical retraction of the

wire. For these calculations, all the equations (thermal, fluid mechanics, electromagnetics) are solved in the metal and the plasma, in association with a level set method used to follow the detachment of the drops and the formation of the deposit (Cadiou 2019). The heat input is calculated from the Joule effect in the plasma. This type of model makes it possible to predict the temperature and velocity field as well as the shape of the deposit from only the operating parameters. Video animations can be seen in Cadiou et al. (2020).

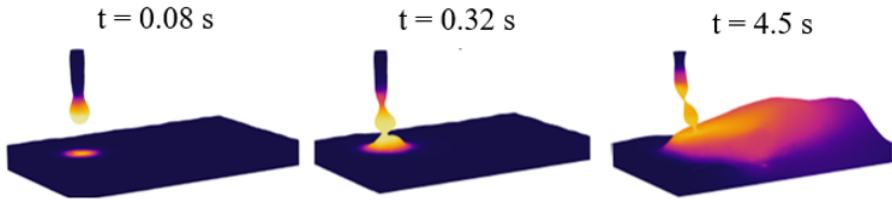


Figure 4.8. *Temperature fields from a magneto-thermo-hydrodynamic model for a WAAM CMT process (Cadiou 2019). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip*

Finally, a third example relates to the selective powder bed melting (L-PBF) process using a laser, simulated by a transient heat transfer and fluid flow problem (Chen et al. 2017; Queva et al. 2020). The powder bed is seen as a continuous medium, the density of which varies abruptly and irreversibly around the melting temperature. As before, the material/gas interface is captured by a level set method. The energy input is distributed either on the surface when the laser hits the molten zone or in the volume when the laser hits the powder bed, due to the multiple reflections occurring in the latter. The very high temperatures mean the effects induced by vaporization (heat flow extraction, recoil pressure) must be considered to explain the high deformation of the surface of the melt-pool, which evolves toward a so-called keyhole capillary (Figure 4.9). Surface tension and the Marangoni effect must also be taken into account. This type of model gives access to the bead shape, which results from the different hydrodynamic effects and solidification. To ensure a sustainable compromise between calculation time, precision and robustness, the modeling is based on a multi-objective adaptive automatic remeshing strategy, with the metric controlling the density and the orientation of the mesh depending on the variables calculated by the solver.

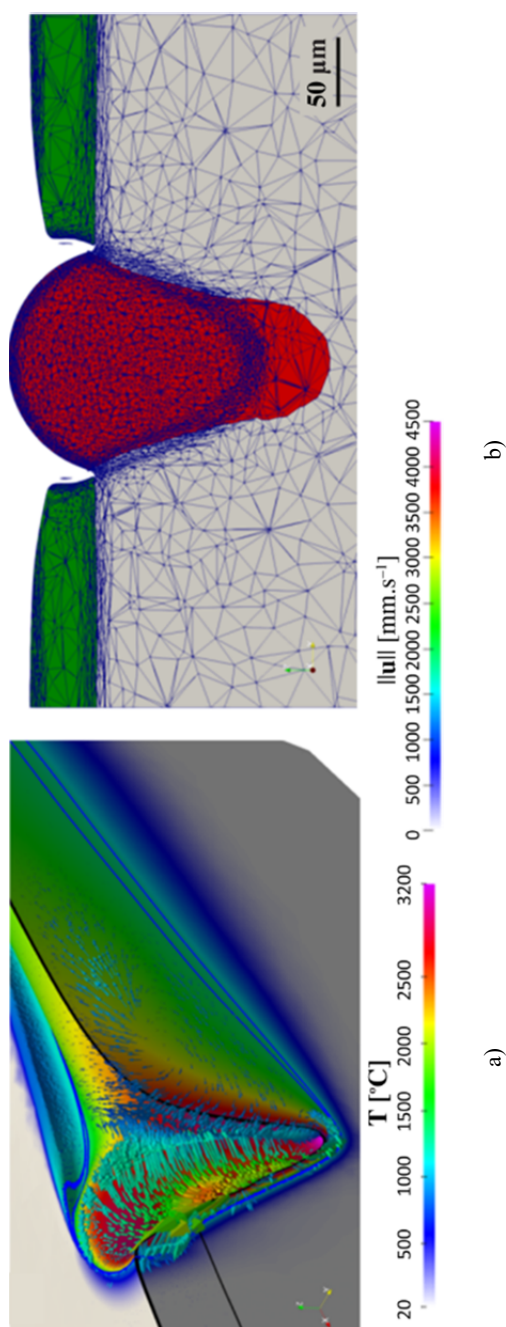


Figure 4.9. Thermo-hydrodynamic simulation of an L-PBF process. (a) Mesh in a cross-section behind the keyhole, with the powder in green and the fused area in red (Queva et al. 2020). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

The simulation of several juxtaposed beads leads to a better understanding of the phenomena causing the enlargement of the melt-pool at the cusp of the laser beam and to a particular surface topography of the deposits in these regions, as illustrated in Figure 4.10.

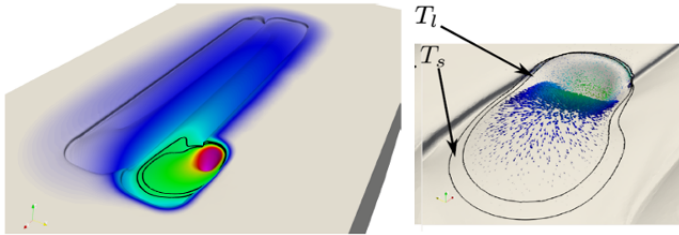


Figure 4.10. Simulation of contiguous bead formation in L-PBF (Queva et al. 2020). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

Finally, the last example deals with the metal vapor jet and the gas flows induced around the melt-pool during the L-PBF process. Figure 4.11 illustrates the velocity field, the pressure field and the concentration of metal vapor which expands in argon, obtained from an axisymmetric 2D multiphysics simulation (Mayi 2019). This type of model is able to deal with both the hydrodynamics of the melt-pool and the expansion dynamics of the metal vapor, with a liquid/gas interface described by the ALE method.

As in the previous example, the surface temperature of the melt-pool exceeds the boiling point of the alloy in question. As a result, the metal vapor expands in the surrounding atmosphere, at a speed of around several hundred meters per second (here up to $400 \text{ m}\cdot\text{s}^{-1}$), or even several thousand meters per second (Li et al. 2020), depending on the material irradiated and the intensity of the incident laser. On the other hand, since the surrounding gas is at rest, the front of the vapor jet progresses in argon much more slowly than the ejection speed (about $80 \text{ m}\cdot\text{s}^{-1}$ in the example given in Figure 4.11). Also, due to the density gradient between the metal vapor and the surrounding atmosphere, the vapor plume assumes a “mushroom” shape, characteristic of the Rayleigh–Taylor instability (Kull 1991). Finally, during expansion, the vapor jet shears the ambient gas on its edges, forming a recirculation loop visible in Figure 4.11. According to Bernoulli’s law, a depression therefore forms near the steam jet, sucking the ambient gas and the surrounding powder, and leading to the denudation phenomenon described in Chapter 3 of this volume.

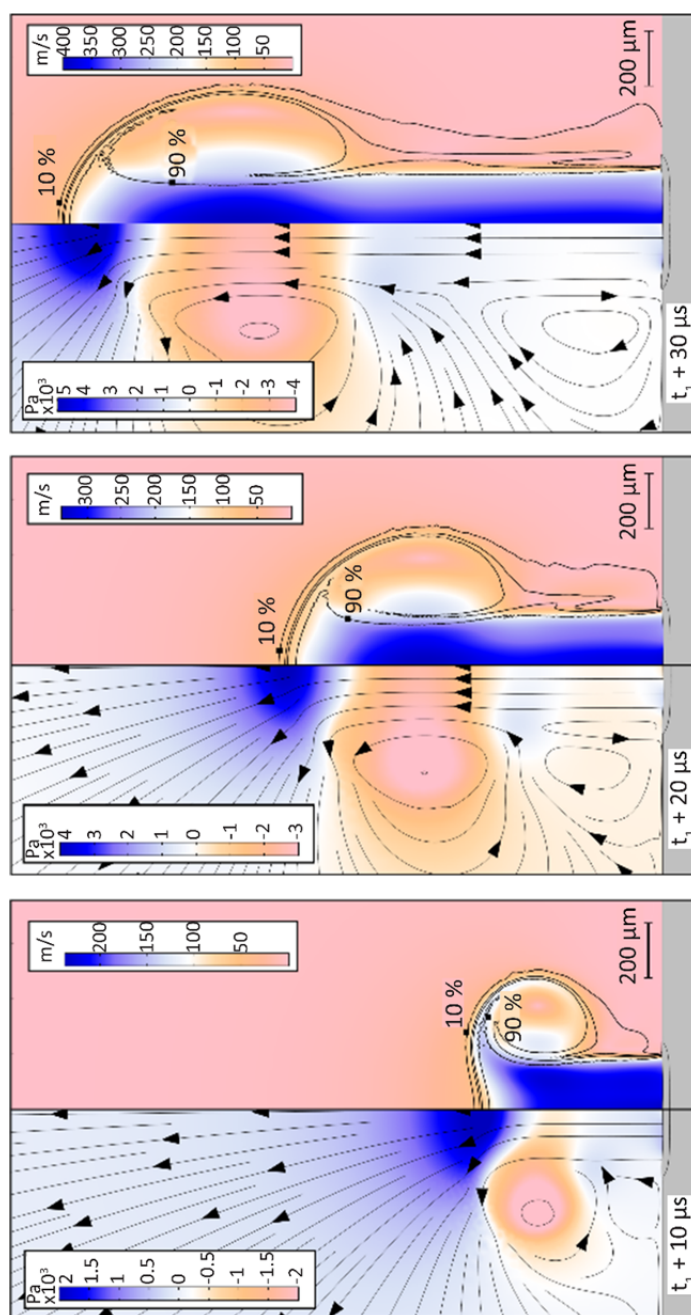


Figure 4.11. Two-dimensional axisymmetric simulation of metal vapor expansion during an L-PBF process. On the left, pressure field and current lines; on the right, velocity field and molar fraction of metal vapor (Mayi 2021). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

Using a numerical model as presented in Figure 4.11 and a simple analytical model of powder particle entrainment, it is possible to predict how the denudation phenomenon will be amplified, or on the contrary attenuated, when the working atmosphere is changed. Figure 4.12 illustrates a typical example.

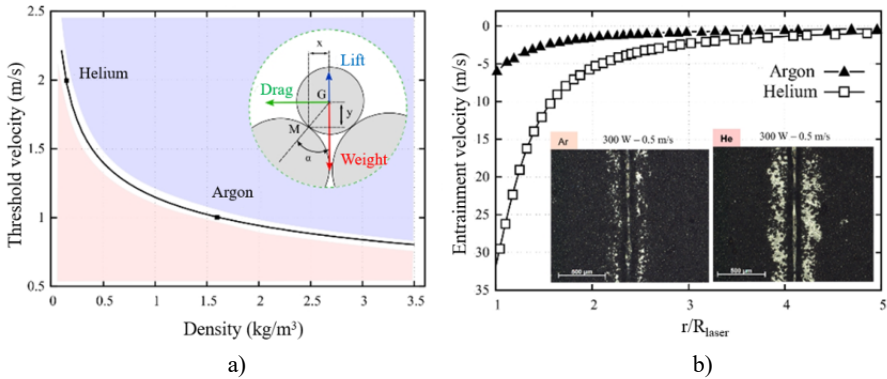


Figure 4.12. a) Particle entrainment threshold as a function of the density of the working gas, calculated from a simplified particle entrainment model (insert on the curve). b) Radial entrainment velocity calculated at the height of the powder bed as a function of the r/R_{laser} ratio (r : distance radial to the axis of the laser beam, R_{laser} : laser radius) (Mayi 2021). The denuded areas inset are taken from Traore et al. (2021). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

Since the kinetic energy of the entrained gas is proportional to the product of its density and the square of its velocity, the lower the density of the gas, the greater the shear rate must be to entrain a powder particle. By taking into account the interaction between the particles at the entrainment threshold, the simple criterion illustrated in Figure 4.12(a) gives a shear rate twice as high, when moving from argon to helium (density divided by 10), to entrain a powder particle (Mayi 2019). On the other hand, at constant laser parameters, the maximum shear rate generated by the plume calculated via digital simulation is six times greater under helium than under argon (Figure 4.12(b)), which largely compensates for the density deficit of helium. This result therefore suggests that under the same parametric conditions, the powder bed is likely to be more efficiently denuded under helium than under argon. Qualitatively, this result is well-verified in various experimental studies (Bidare et al. 2018; Traore et al. 2021).

This last example therefore illustrates two points. First, if a thermo-hydrodynamic simulation is crucial in L-PBF in order to understand how the laser–material interaction parameters condition the stability of the weld pool, it is

just as important to show that changing the shielding gas without changing the laser–material interaction conditions has an indirect effect on the manufacture via the recirculation phenomena around the fusion zone. Second, this example also shows that numerical simulation, even on a very local scale, can be used, coupled with simple analytical approaches, to construct guides that help the machine user select appropriate manufacturing parameters – here the choice of a working gas to minimize denudation.

4.2. Thermomechanical simulation

During and after the additive manufacturing process, the presence of thermomechanical defects can usually be detected, such as cracking and distortion. On a macroscopic scale, high temperature gradients and thermal cycling can produce, in the vicinity of the construction front, cracks with a plane perpendicular to the direction of manufacture (Figure 4.13) (Kempen et al. 2013). Cracks can also appear at the part–substrate junction where the part’s tendency to bend can induce strong tensile stresses in the direction of manufacture. These can be highlighted during the separation of the part from its substrate by machining or wire cutting, which can induce a marked distortion of the part (Figure 4.13) (Jonsson and Krappedal 2018).

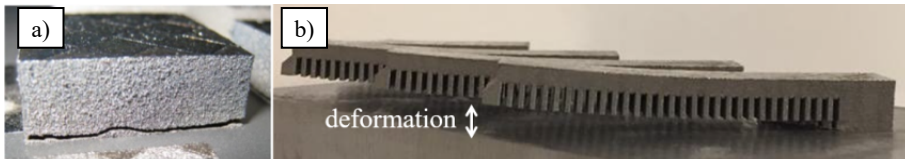


Figure 4.13. (a) Cracked M2 high speed steel (HSS) part (Kempen et al. 2013); (b) deformation of parts after separation from the substrate (Jonsson and Krappedal 2018)

Cracks can also appear in the mass of the constructed part. These cracks start not in the solid state like the previous ones, but at the periphery or at the rear of the fusion zone, when the local stress state puts the metal, which is not yet completely solidified, under tension. This is the hot cracking phenomenon, well-known in welding, which mainly affects metallurgical grain boundaries (Figure 4.14) (Carter et al. 2012).

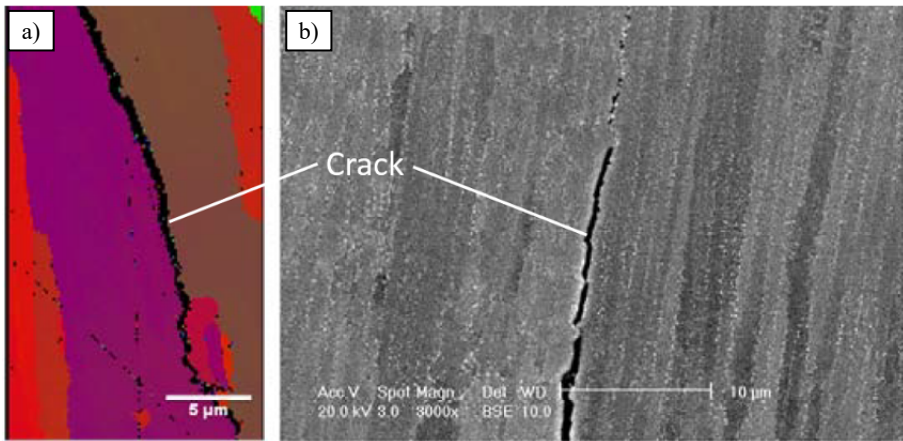


Figure 4.14. a) Electron back scattering diffraction (EBSD) map showing a crack between two grains of a nickel-base superalloy; b) scanning electron micrograph of a similar crack (Carter et al. 2012). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

To understand and avoid all of these defects, it is necessary to develop coupled thermomechanical models, adapted either to the part (distortions, residual stresses), or to the elementary deposits and fusion zones (hot cracking). In this context, it is also necessary to take into account the metallurgical evolution of materials during the modeling of additive manufacturing processes, because of the couplings between thermal, metallurgical evolution and mechanical behavior. The modeling of metallurgical transformations is also part of a larger framework, aimed at predicting the properties of manufactured parts and gradually mastering additive manufacturing processes.

In section 4.2.1, the different numerical simulation methods with regard to a complete part are presented. Depending on its complexity, whether or not to explicitly follow the path of the energy source may be considered in the simulation. When explicit monitoring is not possible, alternative methods are presented. Sections 4.2.2, 4.2.3 and 4.2.4 present the modeling of the metallurgical transformation kinetics and the thermomechanics of the solid. The coupling between the different simulations (thermal, metallurgical, mechanical) is the subject of section 4.2.5. Finally, sections 4.2.6 to 4.2.8 give application examples.

4.2.1. Whole-part simulation: different techniques

The two methods usually called the quiet element and inactive element or element birth methods are the most common. In the quiet element technique, the entire part to be built is meshed at the start of the process. However, the thermal and mechanical properties of each element are artificially lowered until the element is reached by the construction front. The element birth technique involves adding, as the construction progresses, the elements attained in the calculation area, which therefore changes during the process. Although this technique has a more physical meaning, it is difficult to implement in finite element (FE) codes, because the elements must be constantly renumbered and reconnected, thus increasing the computation costs.

These two methods can be used for LMD and WAAM processes for which the material additions are of significant thickness compared to the dimensions of the parts, thus limiting the number of successive deposits. However, a little calculation is enough to realize the impossibility of applying this methodology to an entire part manufactured by L-PBF. For a 10 mm square cube made with a 70 μm diameter laser, 50 μm hatch spacing distance and 40 μm layer thickness, the total scanned length would be 500 m. For a part of a few centimeters in size with a complex geometry, the length of the bead would be tens of kilometers, making it impossible to use digital solutions that follow the source. We therefore proceed differently.

To limit the calculation times, a technique used by many authors involves depositing the material more globally by laser line, by fraction of a layer or even by entire layer deposition. The simulation must then be adapted accordingly, in particular from the point of view of the energy input. This technique is usually implemented with the quiet element or element birth methods (Chiumenti et al. 2017). Note, however, that with these methods, the initial mesh must be structured in successive layers of elements, which constitutes a strong constraint. This is why, since the additions represent low thicknesses, like in L-PBF, Zhang et al (2018b) use the level set method to simulate the progression of the construction front in the mesh of the part: the method allows the monitoring of the material/gas interface. In addition, to increase the precision, the elemental deposition can be reduced to a fraction of a layer. Layer fractions are identified directly from a description of the laser scan plane of the part to be built. Each fraction is heated for a time interval corresponding to the exposure time to the laser beam, then cooled for a time interval equal to the scan time of the layer fraction considered. Conversely, to further reduce the computation time, the elementary deposit can be made up of several layers deposited simultaneously: this super-layer type approach is used by many teams.

4.2.2. Heat transfer resolution

4.2.2.1. Heat transfer equation

Heat transfer is governed by the energy conservation equation:

$$\frac{\partial(\rho h)}{\partial t} + \nabla \cdot (\rho h \vec{u}) - \nabla \cdot (K \nabla T) = S_v \quad [4.26]$$

where ρ is the density, h is the mass enthalpy, $\vec{u}$ is the velocity field, K is the thermal conductivity, T is the temperature and S_v is a source term, which we will come back to later. The mass enthalpy h is related to temperature via the specific heat c_p in the solid regions, or in the powder bed constituting the last layer deposited. In general, in the event of a phase change such as fusion or solidification, the enthalpy of transformation (latent heat) is also involved in this enthalpy–temperature relationship. However, it should be noted, on the one hand, that the characteristic time of melting or solidification (10^{-5} to 10^{-4} s) is very small compared to the heat diffusion in the part, and, on the other hand, the characteristic size of the fusion zone (several 100 μm) is very small compared to that of the part. The excess energy required by fusion is therefore compensated – almost in the same place and at the same time – by the energy release during solidification. Thus, as in welding, for a simulation at the macroscopic scale, it is legitimate to neglect the effects of enthalpy in the energy equation. For simulations on a smaller scale (that of fusion zones or elementary beads), it is of course quite different since the shape of the fusion zones, in particular the part rear, will be a function of the enthalpy and of the kinetics of crystallization characteristic of each alloy and its thermodynamic properties (see section 4.2.3).

Depending on the types of simulation, the energy supply can be affected by means of surface fluxes imposed on the surface of the melt-pool or the powder bed. These flows therefore constitute boundary conditions that the calculated solution must satisfy. When the metal–gas interface is diffuse in the computational mesh, as in the context of a level set formulation, these surface terms are converted into a volume source, by means of multiplication using the numerical Dirac function, to form the term S_v in the second part of equation [4.26]. In the case of a macroscopic approach, with deposition by layer(s) or layer fractions, as described in section 4.2.1, the source term S_v can be distributed throughout the thickness of the layer to express the volume power density associated with the laser/material interaction averaged over the elementary deposit (Zhang et al. 2018a).

4.2.2.2. Thermal properties of the powder bed

When solving the heat equation, convective or radiative losses are to be taken into account at the interfaces. In contrast, the powder bed melting process involves the thermal properties of the powder bed. In equation [4.26], it is appropriate to take, for a porous granular medium, on the one hand, $\rho_{app} = \rho_{ref}(1 - \omega)$, with ρ_{ref} the density of the material constituting the particles and ω the porosity ; and, on the other hand, $h = \int c_p(T)dT$, with $c_p = c_{p,ref}$. The difficulty arises from estimating the apparent conductivity of the medium, K , which is a function of K_{ref} , K_{gas} and ω . Different models from the literature can be used, such as the model of Sih and Barlow (2004).

In the case of the L-PBF process, the powder bed not exposed to the laser plays a certain role in the diffusion of heat in the vicinity of the parts under construction. In numerical simulations, several approaches are then considered. A rough approach is to ignore this aspect by assigning, to the surface of the part, either an adiabatic condition or a heat exchange condition of convection with an arbitrary coefficient. An intermediate approach involves justifying the value of this coefficient by considering the 1D heat diffusion in a powder medium, in the direction normal to the wall of the part. Finally, the most complete approach involves including, and therefore discretizing into FEs, the non-lasered powder in the overall heat transfer solving process.

4.2.3. Metallurgical resolution

In additive manufacturing, the microstructure has a significant influence on mechanical properties at higher scales, in particular via interactions between the different constituents formed (Liu et al. 2015). In addition, the observed microstructures often develop incrementally with rapid thermal cycles, leading to strongly oriented structures at the local scale. Effective modeling strategies therefore must consider the nonlinear nature of behavior and the effect of microstructures, such as by using non-uniform transformation field analysis (NTFA) or proper orthogonal decomposition (POD) (Liu et al. 2015).

More particularly, the modeling of the metallurgical transformations, including solidification and cooling, must make it possible to obtain a relevant estimate of the mechanical constitutive laws of the material and the related effects (evolution of stresses, appearance of cracking defects, etc.). It is thus necessary to determine anisotropic constitutive laws relating to the materials produced by considering the

effect of the orientations, sizes and morphologies of the grains on this mechanical behavior. Kok et al. (2018) have analyzed the causes of anisotropies and microstructural heterogeneities encountered in AM processes, as well as their mechanical effects on the performance of parts. The evolution of the phase fractions formed during the various deposits and their inhomogeneity during construction are reported, in particular for Ti-6Al-4V alloys for which they are particularly noticeable. We focus here more specifically on the predictions of phase transformations and changes in grain structures from recent models.

4.2.3.1. *Phase transformations*

Additive manufacturing processes lead to rapid phase transformations. Thus, in L-PBF processes applied to metals, cooling rates of between 10^4 and 10^6 K·s⁻¹ are commonly encountered (Gokuldoss et al. 2017). The high energy densities and the small dimensions of the melt-pool, but also the high laser speeds, can lead to metallurgical transformations out of thermodynamic equilibrium or to the formation of very fine microstructures, compared to classic production processes. In addition, the rapid heating/cooling cycles have an amplitude dependent on the distance from the heat source. These particularities complicate the prediction of the observed metallurgical transformations, but are necessary to estimate the final properties of the parts. Thus, these cumulative transformations condition the nature of the phases observed after elaboration and the associated constitutive laws. In particular, the morphology of the structures formed, and their complexity, can often allow an improvement in the mechanical strength.

The microstructural modeling of materials developed in additive manufacturing however remains a particular challenge that should guide the future of simulation activities (Smith et al. 2016a). The microstructural characteristics affected correspond as much to the nature and importance of the phases formed and the morphological characteristics of the grains as to the development of final porosities and inclusions (Smith et al. 2016a, 2016b). In particular, it should be noted that few numerical approaches allow us to focus on phase transformations over all construction cycles. Nevertheless, Smith et al. note that the Computer Coupling of Phase Diagrams and Thermochemistry (CALPHAD) approaches are very promising, integrating the thermodynamic properties of the different phases formed into the description of material transformations. The predicted behavior makes it possible to consider both the phase equilibria and the properties of the interfacial equilibria in estimating the growth kinetics. Developments coupling CALPHAD approaches, the heat transfer problem and FE modeling, at the macroscopic scale, have recently emerged (Smith et al. 2016b), allowing the prediction of transformations in complex thermal

cycles. However, the authors highlight the need for relevant experimental calibrations to better estimate the thermal kinetics.

Different authors have focused on the prediction of metallurgical transformations, on the basis of precise and uncoupled knowledge of the thermal evolutions of parts. For Ti-6Al-4V titanium alloys, Salsi et al. (2018) have developed a detailed model allowing this monitoring of transformations by taking up, developing and enriching an approach already present in the literature. The latter is based on the use of time–temperature–transformation (TTT) diagrams specific to each alloy, giving the start and end times of transformation of the phase changes studied. Over the transformation interval, it is usually considered that the phase fractions formed, f^φ , follow a temporal evolution, for an isothermal transformation, in the form of a Johnson–Mehl–Avrami–Kolmogorov (JMAK) law, which, neglecting the incubation time, can be formulated as:

$$f^\varphi = X^0 X^{eq} (1 - e^{-k t^n}) \quad [4.27]$$

with X^0 being the initial fraction of the transformed phase available and X^{eq} the phase fraction obtained at thermodynamic equilibrium. The parameters k and n , functions of temperature, are estimated with measurements of fractions associated with the critical thresholds of transformation. In addition, non-diffusive martensitic transformations are usually predicted by Koistinen–Marburger laws in the form:

$$f^M = X_1^0 (1 - e^{-k (T^M - T)}) \quad [4.28]$$

In this context, the transformations are only a function of the local temperature with regard to the temperature at the start of martensitic transformation, T^M . Relations are added to take into account the dissolution of the phases under heating, according to their nature (Figure 4.15) and specific kinetics.

The preceding relations must thus be applied throughout the thermal cycle $T(t)$ as a succession of consecutive infinitesimal steps, each of these steps being deemed isothermal, according to the additivity principle of transformations. In this context, however, it is necessary to correctly locate the phases present on each isotherm, which is achieved through a conventional principle of calculating equivalent transformation time. Figure 4.16 thus illustrates the results obtained from the prediction of the martensitic fraction formed in the vertical plane of a section of a Ti-6Al-4V part lasered under two different conditions (the distance between different passes) for an L-PBF process. We thus find strongly differentiated profiles, with higher martensitic phase fractions in the second work situation.

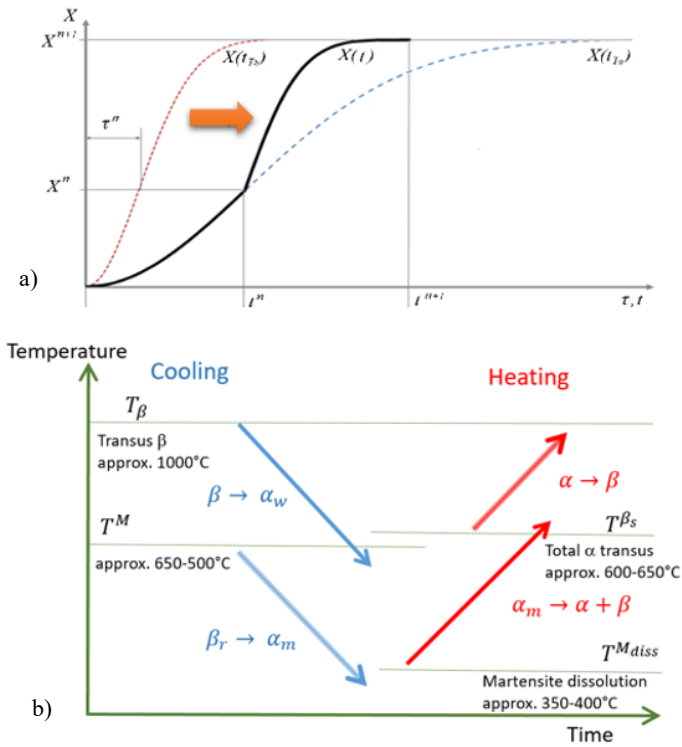


Figure 4.15. a) Evolution of phase fractions formed; b) metallurgical transformations considered for the alloy Ti-6Al-4V. From Salsi et al. (2018). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

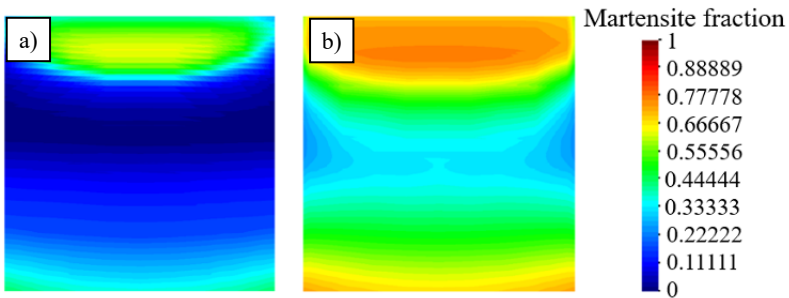


Figure 4.16. Final fraction of martensite ($P_0 = 375 \text{ W}$, $V_0 = 1,029 \text{ mm} \cdot \text{s}^{-1}$). Distances between passes at a) 0.12 mm and b) 0.18 mm. From Salsi et al. (2018). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

4.2.3.2. Grain structures

It is also necessary to examine the developments of the grain structures formed under these solidification conditions. Different models are present in the literature. However, the cellular automaton (CA) approach, coupled with the solving of macroscopic conservation equations, has several advantages. This approach makes it possible, on a spatial scale adapted to the process, to follow the stages of grain nucleation by integrating the mechanisms of growth competitions. The sizes, crystallographic orientations and morphology of the solidification structures are, in this approach, determined at the scale of the beads of material deposited. Thus, for an E-PBF process, Rai et al. (2016) have developed a coupled lattice Boltzmann (LB)–cellular automaton (CA) method to estimate the evolution of grain textures formed in 2D. This approach integrates multiphysical modeling of heat transfer (electron beam, vaporization, etc.) and fluid dynamics (flows, wetting, etc.) to determine the associated microstructural evolutions. The thermodynamic and hydrodynamic resolutions are based on the LB method, with a free surface. Figure 4.17(a) illustrates the cell capture mechanism proposed to reproduce grain growth. Simple, square forms of growth are thus associated with the different cells. Their kinetics are determined by local supercooling, deduced from heat transfer resolution. The development of these forms allows the progressive recovery, step by step, of the cells which have remained free, with which the orientations of the capturing cells are associated. This mechanism mimics the process of grain growth, and, on a larger spatial scale, gives a representation of the developed microstructure. The CA approach also makes it possible to simulate the pre-existing solidified microstructure, on which the powder is deposited. The arrangement of the grains on the powder bed is obtained by a specific algorithm mimicking the consolidation of the bed under the effect of gravity and also integrating the contacts between grains (Figure 4.17(b)).

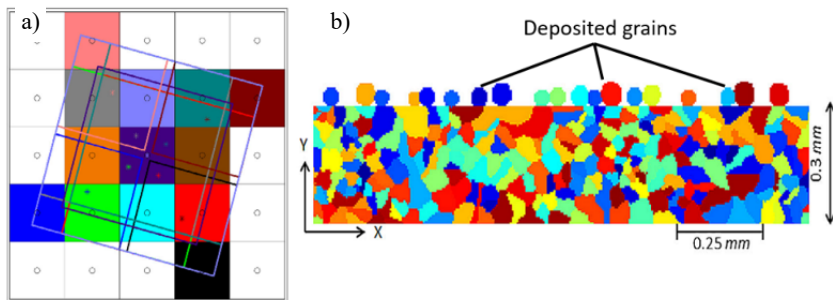


Figure 4.17. a) Growth mechanism showing the coated cells and associated growth forms. b) System illustrating the basic microstructure and the grains deposited to form the upper layer (Rai et al. 2016). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

The grain structure formed after different passes can thus be simulated, highlighting the mechanisms of epitaxial growth and selection of grain orientation. Competitive structural growth thus leads to the persistence of large grains and restricted orientations. Figure 4.18 illustrates the final grain structure of an Inconel alloy and the associated orientation distribution after building a 1.7 mm high wall for an alternating sweep strategy. The grain structure appears relatively symmetrical, with good penetration of the powder on both sides, with regard to the travel conditions. A single direction of travel leads, conversely, to a preferred orientation of the grains in the upper layer of $\sim -36^\circ$ and $\sim +28^\circ$ (Rai et al. 2016) for paths oriented to the left and the right, respectively, in connection with the temperature fields and the shape of the melt-pool. The 2D character of the simulations, linked to the long computation time of the approach (the LBCA method), allows the appropriate study of the influence of process parameters or construction conditions.

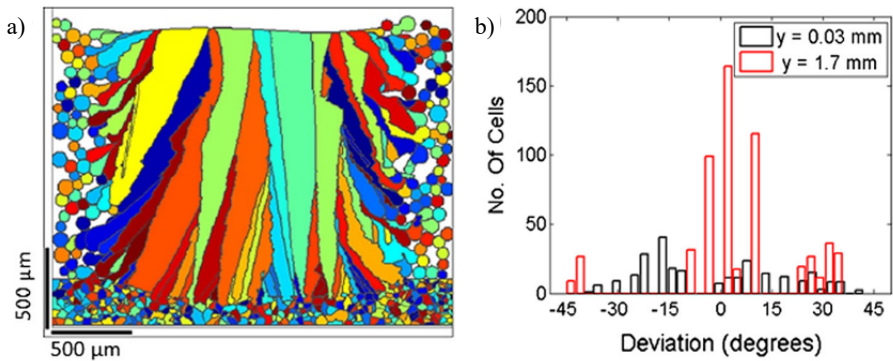


Figure 4.18. a) Microstructural evolution and b) distribution of grain orientations at different heights ($P_0 = 594 \text{ W}$, $V_0 = 2.2 \text{ m}\cdot\text{s}^{-1}$), according to Rai et al. (2016). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

4.2.4. Mechanical resolution

4.2.4.1. Equilibrium equation

The Cauchy stress tensor denoted σ satisfies the equilibrium equation, which is most often solved using the FE method and is written, neglecting the effects of inertia and gravity:

$$\nabla \cdot \sigma = 0 \quad [4.29]$$

4.2.4.2. Mechanical constitutive law

Given that the material in the solid state must be taken into account in a very wide temperature range, from ambient temperature to melting point, its behavior must be modeled using an elasto-viscoplastic constitutive model. Models of perfect plasticity or elastoplasticity are not appropriate. Among the many possible variations, the behavior can be written as follows:

$$\dot{\epsilon} = \dot{\epsilon}^{el} + \dot{\epsilon}^{vp} + \dot{\epsilon}^{th} \quad [4.30]$$

$$\dot{\epsilon}^{el} = \frac{1+\nu}{E} \dot{\sigma} - \frac{\nu}{E} \text{tr}(\dot{\sigma}) \mathbf{I} \quad [4.31]$$

$$\dot{\epsilon}^{vp} = \frac{3}{2\bar{\sigma}} \left[\frac{\bar{\sigma} - (\sigma_Y + R(\bar{\epsilon}))}{k_{vp}(\bar{\epsilon})} \right]_+^{\frac{1}{m}} \mathbf{s} \quad [4.32]$$

$$\dot{\epsilon}^{th} = -\frac{1}{3\rho} \frac{d\rho}{dt} \mathbf{I} \quad [4.33]$$

The strain rate tensor $\dot{\epsilon}$ is divided into an elastic part $\dot{\epsilon}^{el}$, a viscoplastic part $\dot{\epsilon}^{vp}$ and a thermal part $\dot{\epsilon}^{th}$ reflecting thermal expansion. The expression of $\dot{\epsilon}^{el}$ is given by Hooke's hypoelastic law (E Young's modulus, ν Poisson coefficient). The expression of $\dot{\epsilon}^{vp}$ refers to the normality rule: the viscoplastic strain rate derives from a potential (not expressed here, a function of the stress tensor) and is proportional to the stress deviator $\mathbf{s}$. The elastic limit, below which no viscoplastic deformation occurs, is denoted $\sigma_Y + R(\bar{\epsilon})$: the variable $\bar{\sigma}$ denotes the equivalent von Mises stress and the expression in square brackets in [4.32] is reduced to zero when negative. This elastic limit depends on an isotropic strain hardening function R of variable expression depending on the cumulative viscoplastic deformation $\bar{\epsilon}$. Consequently, we use the following relation:

$$\bar{\sigma} = \sigma_Y + R(\bar{\epsilon}) + k_{vp}(\bar{\epsilon}) \dot{\epsilon}^m \quad [4.34]$$

The characteristic parameters of the material depend on the temperature and must be characterized: E, ν, σ_Y and m and the function parameters R and k_{vp} . In addition to isotropic hardening, it may be relevant in AM to consider kinematic hardening, in other words the displacement of the elasticity domain in the stress space. Indeed, the material can undergo a succession of load reversals due to thermal cycling during the construction process. The displacement in space of the stresses is characterized by an internal stress tensor, denoted χ . Equations [4.30] to [4.34] can be conserved, with equation [4.32] being modified, so that, on the one hand, $\bar{\sigma}$ is replaced by the same function expressed this time for the tensor $\sigma - \chi$: $\sqrt{(3/2)(\sigma - \chi)':(\sigma - \chi)'}$, and, on the other hand, $\mathbf{s} = \sigma'$ is replaced by $(\sigma - \chi)'$,

with $(\cdot)'$ being the deviation of a tensor. The evolution of the tensor χ can be calculated by an expression of the following type:

$$\dot{\chi} = \frac{2}{3} C \dot{\epsilon}^{vp} - \gamma \dot{\epsilon} \chi \quad [4.35]$$

with C and γ being the parameters of the material considered.

4.2.4.3. Processing of nonlinearities: inherent strain method

Taking into account the elasto-viscoplastic behavior, it can be shown that the resolution of the FE problem, based on a weak formulation of equation [4.29], is nonlinear according to the discretized displacement or velocity field. Thus, one must proceed by Newton–Raphson iterations to converge toward the solution. In an attempt to reduce the computation time by avoiding nonlinearities, the inherent strain method has been proposed (Yuan and Ueda 1996). Resulting from the simulation of welding processes, it is based on the observation that the viscoplastic deformations are concentrated around the energy supply zone, with a relatively constant spatial distribution all along a weld bead. Thus, if we assume that this distribution is known (using a simulation carried out beforehand for example), we can modify equation [4.30] as follows:

$$\epsilon = \epsilon^{el} + \epsilon^* + \epsilon^{th} \quad [4.36]$$

with ϵ^* being the inherent deformation, which therefore does not depend on the velocity field to be determined. The resolution of the problem then becomes linear, requiring only a single resolution, hence its success in additive manufacturing. However, already questionable in single-pass welding, it poses a certain number of questions in additive manufacturing. Variants have been proposed, such as that involving the consideration of the decomposition of the strain rates at each time step of a manufacture, with an inherent strain rate $\dot{\epsilon}^*$ whose field is known:

$$\dot{\epsilon} = \dot{\epsilon}^{el} + \dot{\epsilon}^* + \dot{\epsilon}^{th} \quad [4.37]$$

4.2.5. Coupling

For an unsteady approach, the thermal, metallurgical and mechanical coupling must be carried out at each time step, most often in a so-called “weak” way; that is to say that the three problems are solved successively before advancing to the next time increment. The coupling between thermal and metallurgical resolutions is discussed in section 4.2.3. Regarding the coupling between metallurgical and mechanical resolutions, it is linked to the behavior of a multiphase material that

depends on the local distribution, the organization of the different phases and their specific characteristics. Knowing the phase change and precipitation kinetics, it is possible, by means of homogenization models, to deduce the average mechanical behavior of the material. In addition, these phase changes may be accompanied by deformation, either by volume due to differences in density between phases, or deviatoric due to the shear accompanying a transformation (such as martensitic transformation for example). The return coupling, from mechanics to metallurgy, can consist of a shift in the metallurgical transformation kinetics as a function of the local stress state. The coupling between thermal and mechanical resolution phenomena results from the dependence of the mechanical properties with the temperature and the thermal expansion terms. The return coupling, from mechanics to thermal properties, is negligible in solid mechanics.

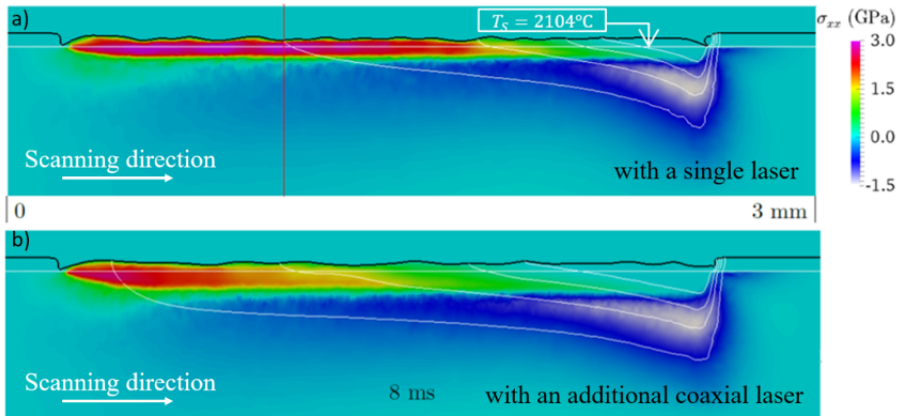


Figure 4.19. Axial stresses with a) a single laser and b) an additional coaxial laser. The isotherms are in white (500, 1,000, 1,500 and 2,104°C, solidus). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

4.2.6. Application at the mesoscopic scale: local manufacturing stresses

It is interesting to calculate the stresses that form in the bead deposited when it is subjected to the heat source, as well as in its immediate environment, in particular to estimate the risk of cracking during the process. Figure 4.19 shows the advantage of using two coaxial lasers in the case of an L-PBF process applied to a ceramic (alumina) (Chen et al. 2017). A first laser ensures the fusion as usual. The second laser, less powerful and defocused, preheats the powder bed in front of the first

laser and keeps the bead warm to relieve stress behind. This combined action significantly reduces the stresses (axial in this case) in the bead and its vicinity behind the melt-pool, creating a more favorable situation for the avoidance of cracking.

4.2.7. Application at the macroscopic scale

4.2.7.1. Residual stresses and distortion

The simulations presented in this section were carried out using the software Cimlib developed at CEMEF (Zhang et al. 2018a, 2018b). The first example is the manufacture of a nickel-based superalloy propeller (Inconel 718) with the L-PBF process. On the basis of a coupled thermomechanical calculation, after cooling to ambient temperature, the distributions of the residual stresses are given first in the part/substrate assembly and second in the part alone after separation from the substrate (Figure 4.20). Since distributions of σ_{xx} and σ_{yy} are similar in horizontal directions, only σ_{xx} is presented. The residual tensile stresses are located mainly in the substrate and at the bottom of the constructed part (Figure 4.20(a)), while the compressive stresses are found in the middle of the blades. When the substrate is removed, the stress at the bottom of the part is released to a lower level. Regarding the residual stress in the vertical direction z , before separation of the substrate, the maximum stress is in the transition zone between the part and the substrate. After removal of the substrate, the stresses are considerably reduced in this region of the part, while the stresses in the upper regions undergo negligible changes. The distortion of the part after cooling to ambient temperature is given in Figures 4.20(e) and (f). We represent a displacement field reflecting the geometric difference between the configuration obtained and that resulting from the CAD of the part. The horizontal component u_x was studied before and after removal of the substrate. The maximum and minimum values are approximately 0.3 mm and -0.3 mm, respectively (Figure 4.20(e)), and are located at the upper end of the blades. Distorsions in the vertical direction are of comparable values to those obtained for the horizontal component (Figure 4.20(f)).

In order to validate thermomechanical calculations, researchers from Fraunhofer Berlin (Biegler et al. 2018) measured displacement fields *in situ* during the LMD construction of a wall made of 316L steel. This 18.6 mm high wall consisted of 30 layers. After depositing the first 20, a speckle of paint was applied to one of the large faces, making it possible, using image correlation, to measure the displacement field during the deposition of the last 10 layers. The numerical macrosimulation developed at CEMEF was applied to the entire construction of this wall, making it possible to compare the experimental and calculated displacement fields (Figure 4.21).

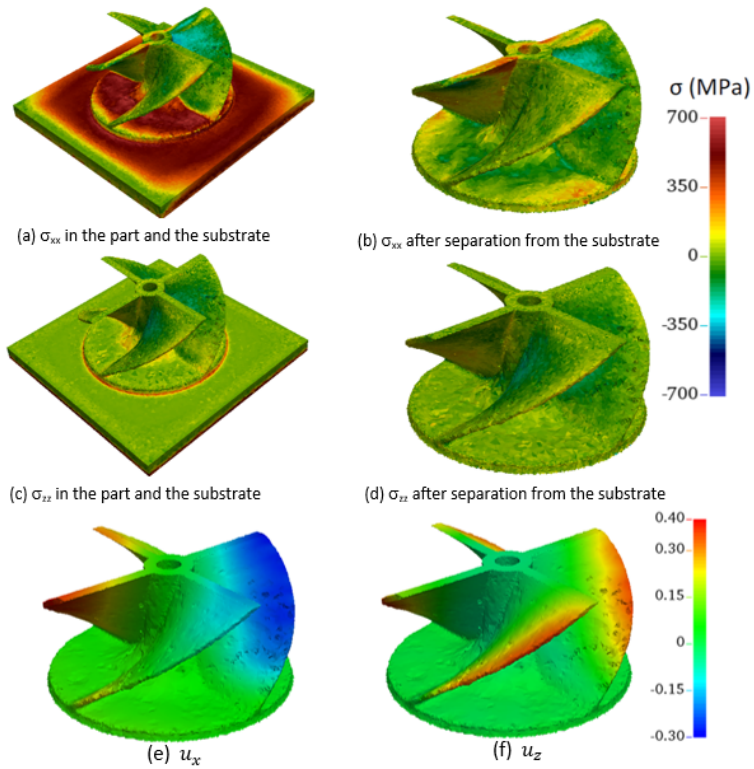


Figure 4.20. Residual stresses after cooling to ambient temperature (a and c) and after separation from the substrate (b and d). Distortion after cooling to ambient temperature and separation from the substrate (e and f). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

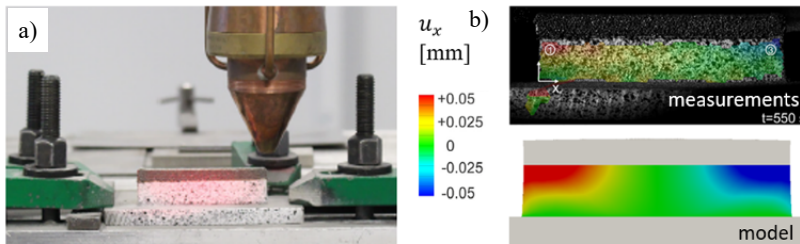


Figure 4.21. a) Experimental set-up with patterned sample, according to Biegler et al. (2018); b) comparison of the horizontal displacement field: measured and calculated. For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

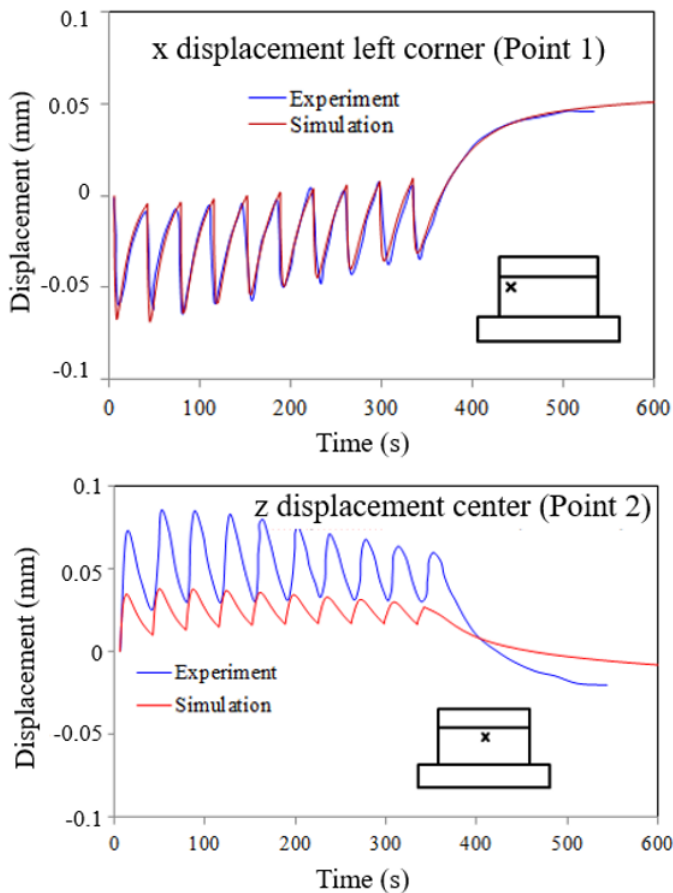


Figure 4.22. Comparison of horizontal and vertical displacements at two different points of a 20-layer wall during the deposition of 10 additional layers. For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

Figure 4.22 shows excellent agreement for the horizontal displacements during the last 10 layers. In each cycle, there is a very rapid displacement during heating, followed by a slower displacement in the opposite direction related to cooling and relaxation of the stresses exerted during heating. On the other hand, the vertical displacements are underestimated in the simulation, which could be explained by the failure to take into account the anisotropy of the material due to the orientation of the microstructure according to the direction of manufacture.

4.2.7.2. Application of the inherent strain method

We present here an application example, taken from Liang et al. (2018), of the inherent strain method described in section 4.2.4.3. Average inherent strain values are taken from a small-scale simulation and applied in the macroscopic simulation of a rectangular-contour deposit consisting of 10 layers (Figure 4.23). The average inherent strain tensor for each deposited layer is applied evenly throughout the layer. The figure presents the vertical residual strain profile and shows a good agreement between complete simulation and simulation using inherent strain.

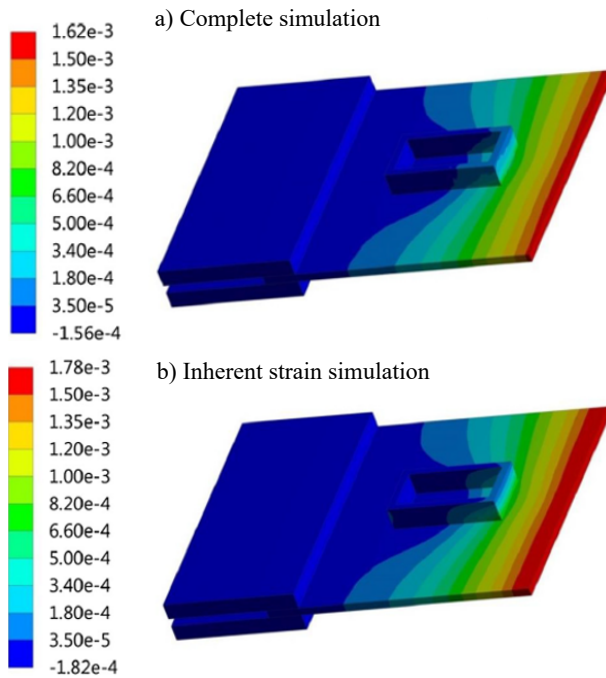


Figure 4.23. Vertical residual distortion (unit: m) calculated for a rectangular deposit with 10 layers, according to Liang et al. (2018). For a color version of this figure, see www.iste.co.uk/peyre/alloys1.zip

It should also be noted that research must be done to reduce the computation times of such models, in particular if one wishes to predict distortions during additive manufacturing of large structures. In the work of Nain (2021), different solutions are examined to reduce calculation times, such as the choice of the heat source and its parameters or numerical methods to process the input of material. The objective is also to implement a promising method of deformation compensation

allowing deformations to be taken into account in the definition of a new digital CAD model (counter-deformations), as shown by Biegler et al. (2020).

4.2.8. Software and calculation codes dedicated to additive manufacturing

There is a fairly wide range of commercial software available for modeling additive manufacturing processes. In almost all cases, this involves the thermomechanics of the whole part, primarily to predict distortions and residual stresses. We give a few examples below, without claiming to be exhaustive. As these software packages are continuously under development, we do not give explicit details on their digital solutions and capabilities; the interested reader is invited to inquire by consulting the referenced sites. Some of these packages also make it possible to tackle thermo-hydrodynamic aspects, such as ESI version 2020 and Flow 3D version 2020. It should be noted that this software can be supplemented, in software suites, by modules aimed at the design of parts for additive fabrication (topological optimization, definition of construction orientation, design of supports):

- ANSYS, specific modules from the company 3DSIM;
- AUTODESK, Netfabb software;
- Dassault Systèmes, 3DEXperience platform;
- ESI, software solution for additive manufacturing;
- Virfac;
- MSC Software, Simufact Additive software;
- Flow 3D AM, computational fluid dynamics software for additive manufacturing.

4.3. Conclusion

This chapter presented the elements necessary for the simulation of the physical phenomena present during additive manufacturing processes for metal parts. Through various examples, the relevance of numerical simulation tools, capable of predicting surface defects, porosities and risks of cracking, has been shown. Multiphysics models also allow a better understanding of the relationships between the operating parameters and the final state of the part and thus a better understanding of the origin of defects. These models are starting to provide guidance for construction strategies to limit the appearance of these defects. In order to meet industrial expectations, efforts must continue to improve the reliability of these

models and reduce computation times. These elements are essential to simulate the complete construction of potentially large parts. Multiscale approaches, capable of managing the coupling between micro-, meso- and macro-scales, remain promising avenues for reducing computation times while guaranteeing the predictive nature of such models. Note that these models can also be coupled with analytical models, which have the advantage of providing much shorter computation times. The development of model reduction algorithms is also a promising way to produce simulations at a lower cost or to provide numerical charts that can be used more easily by technical operators. These digital developments must also be based on extensive experimental validations in order to ensure the reliability of the models.

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Conclusion

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Metal additive manufacturing (AM), approached from the angle of materials and processes as described in this book, calls on a wide range of related scientific topics, including the control of the raw material, the energy input/supply and the physics of the associated processes, to arrive at the microstructures and the materials properties of the manufactured parts. This work, divided into two distinct volumes (Volume 1 dealing with metallic AM processes, raw materials, process physics and numerical simulation; Volume 2 dealing with microstructures, post-processing and materials properties), details each of these different elements in the overall control of the different processes, with applications to different categories of materials, for very specific material properties. On the other hand, we deliberately chose not to talk about innovation and design aspects, which form a topic in their own right and are of course of prime importance in the overall control of the additive manufacturing value chain.

To our knowledge, this book, originally published in French, was the first of its kind in the French language and we hope that it will be useful to as many people as possible, and first and foremost to the growing number of users of these new manufacturing processes.

Finally, due to the very large amount of information produced on the subject and the rapid evolution of technologies, such a document is not itself an exhaustive contribution, but rather a snapshot taken at the beginning of 2021 of what seemed necessary to us to know about this topic.

Additive Manufacturing of Metal Alloys 1,
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Abbreviations

Acronyms

AFM	Abrasive flow machining
ALE	Arbitrary Lagrangian–Eulerian
AM	Additive manufacturing
CA	Cellular automata
CAD	Computer-aided design
CIN	Cylindrical involution normal
CMT	Cold metal transfer
CSAM	Cold spray additive manufacturing
DED	Direct energy deposition; includes WAAM and LMD processes
EBM	Electron beam melting
EBSD	Electron back scattering diffraction
EDX	Energy-dispersive X-ray spectroscopy
E-PBF	Electron powder bed fusion (also known as EBM)
FE	Finite elements

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GED	E_v^G = global energy density per volume manufactured ($\text{J}\cdot\text{m}^{-3}$)
HAZ	Heat-affected zone
HIP	Hot isostatic pressing
HSS	High-speed steel
HV ₀	Vickers hardness
IR	Infrared
JMAK	Johnson–Mehl–Avrami–Kolmogorov
LB	Lattice Boltzmann
LBCA	Lattice Boltzmann–cellular automata
LMD	Laser metal deposition
L-PBF	Laser powder bed fusion
MBJ	Metal binder jetting
MIM	Metal injection molding
MMP	Micro-machining process
MP	Melt-pool
MZ	Melting zone
PeP	Plasma electrolytic polishing
POD	Proper orthogonal decomposition
SEM	Scanning electron microscopy
SLS	Selective laser sintering
TEM	Transmission electron microscopy

TTT	Time–temperature–transformation
UHP	Ultra-high pressure (for waterjet technology)
VED	E_v = volume energy density ($\text{J}\cdot\text{m}^{-3}$)
WAAM	Wire arc additive manufacturing
WLAM	Wire laser additive manufacturing

Latin abbreviations

A	Absorptivity
$A\%$	Ductility, also known as fracture strain (mm/mm)
A_f, A_r, B_s, C_s	Goldak source distribution parameters
A_m	Magnetic vector potential ($\text{T}\cdot\text{m}$)
a_{th}	Thermal diffusivity ($\text{m}^2\cdot\text{s}^{-1}$)
B	Magnetic field (T)
Bo	Bond number
C	Kinematic hardening modulus
C_D	Drag coefficient
C_p	Specific heat ($\text{J}\cdot\text{kg}^{-1}\cdot\text{K}^{-1}$)
D_m	Mass flow rate in DED ($\text{kg}\cdot\text{s}^{-1}$)
D_m^*	Surface mass flow rate in DED ($\text{kg}\cdot\text{s}^{-1}\cdot\text{m}^{-2}$)
D_0	Beam diameter (m)
D_v	Manufacturing volume flow rate ($\text{m}^3\cdot\text{s}^{-1}$)
D_v^{crit}	Production volume flow rate in L-PBF ($\text{m}^3\cdot\text{s}^{-1}$)

d_{50}	Median powder diameter (m)
d_p	Particle diameter (m)
d_{wir}	Wire diameter (m)
E	Young's modulus (Pa)
E_{elec}	Electric field ($V \cdot m^{-1}$)
e_v	Hatch spacing = no scanning (m)
E_v	Volume energy density ($J \cdot m^{-3}$) = I_0/V_0
E_{I^g}	Global energy per built volume (L-PBF) ($J \cdot m^{-3}$)
E_l	Linear energy density ($J \cdot m^{-1}$)
E_0	Incident energy of the electron beam (J)
e, e_{app}	Beam width and apparent width (m)
e_s	Heat source thickness (m)
$\vec{e}_z$	Vertical unit vector
f^φ	Fraction of metallurgical phases
$\vec{f}_v, \vec{f}_m$	Surface tension and Marangoni forces ($N \cdot m^{-2}$)
$\vec{f}_f, \vec{f}_l$	Buoyancy and Lorentz forces ($N \cdot m^{-3}$)
F_{col}	Collimation distance (m)
F_{foc}	Focus distance (m)
G	Thermal gradient ($K \cdot m^{-1}$)
g	Gravity acceleration ($m \cdot s^{-2}$)
G^l, G^α	Free enthalpies (liquid, phase a) (J)

H_{FZ}	Height of melt-pool and/or L-PBF beads (m)
h	Mass enthalpy ($\text{J}\cdot\text{kg}^{-1}$)
h_c	Heat transfer coefficient ($\text{W}\cdot\text{m}^{-2}\cdot\text{K}^{-1}$)
h_{dil}	Dilution height (m)
I	Current, intensity (A)
I_d	Identity matrix
I_0	Power per unit density ($\text{W}\cdot\text{m}^{-2}$)
I_{FE}	Primary current of the electron beam (A)
j	Electric current density ($\text{J}\cdot\text{m}^{-2}$)
K	Thermal conductivity ($\text{W}\cdot\text{m}^{-1}\cdot\text{K}^{-1}$)
K_z	Involution factor (-)
k	Concentration factor (-)
k^α	Partition coefficient (solid α /liquid)
La	Laplace number
L_F	Enthalpy of fusion ($\text{J}\cdot\text{kg}^{-1}$)
L_{ZF}	Length of the fusion zone (m)
L_v	Enthalpy of vaporization ($\text{J}\cdot\text{kg}^{-1}$)
l_{ZF}	Width of melting zone (m)
M	Atomic mass (kg)
m, n	Parameters of the function $g(\text{Pe})$ (analytical welding model) $m^{1\alpha} =$ slope of the liquidus (K)
N_{jet}	Constriction coefficient of the powder jet in DED (-)

N_p	Particle density in the DED powder jet (m^{-3})
n_r	Complex refractive index
P_0	Laser power (W)
Pe_w	Péclet number
P_{rec}	Recoil pressure (Pa)
P_{sat}	Saturating vapor pressure (Pa)
p	Pressure field (Pa)
q_m	Mass flow rate ($\text{kg}\cdot\text{s}^{-1}$)
r_{jet}	Powder jet radius in DED (m)
r	Radial coordinate (m)
r_{jet}	Powder jet radius (m)
R	$8.314 \text{ J}\cdot\text{mol}^{-1}\text{K}^{-1}$ = ideal gas constant
R	Fatigue load ratio ($\sigma_{\min}/\sigma_{\max}$)
Ra	Linear roughness, average height measured on a profile
Re_p	Reynolds number of the particle
R_0	Reflectivity of laser radiation
R_{grain}	Reflectivity of the solid powder grain
R_d	Radius of dendrite tip (m)
R_{ec}	Isotropic hardening modulus
R_{laser}	Laser beam radius (m)
$Rp_{0,2}$	Conventional yield stress (Pa)

Rm	Ultimate tensile stress (Pa)
Sa	Surface roughness, average height measured on a surface
Sdr	Developed interfacial area ratio: percentage of definition area's additional surface area contributed by the texture as compared to the planar definition area
S_p	Cross-section of a powder grain (m^2)
S_{ray}	Radiation loss ($W \cdot m^{-3}$)
S_v	Source term ($W \cdot m^{-3}$)
S_{wir}	Wire cross-section in WAAM and WLAM (m^2)
Sz	Maximum roughness, maximum height measured on a surface
S_0	Laser beam area (m^2)
T	Temperature (K)
T_0	Initial temperature, preheating temperature (powder bed) (K)
T_d^{al}	Dendrite peak temperature (K)
$T_E^{\alpha\beta l}$	Eutectic temperature (K)
T_F	Melting/fusion temperature (K)
T_L^{al}	Liquidus temperature (K)
T_S^{al}	Solidus temperature (K)
T^M	Temperature at the start of martensitic transformation (K)
T_v	Vaporization temperature (K)
$\dot{T}$	Cooling rate ($K \cdot s^{-1}$)
T_∞	Temperature of the surrounding environment (K)
T_β	Temperature of transus b (titanium) (K)

t	Time (s)
t_{int}	Laser–matter interaction time = D_0/V_0 (s)
U	Arc voltage (V)
$U, u, \vec{u}$	Fluid velocity ($\text{m}\cdot\text{s}^{-1}$)
$\vec{u}$	Velocity field vector ($\text{m}\cdot\text{s}^{-1}$)
V	Electric potential (V)
V_f	Wire velocity (WAAM and WLAM) ($\text{m}\cdot\text{s}^{-1}$)
V_{FE}	Average penetration volume of the electron beam (m^3)
V_s	Solidification rate ($\text{m}\cdot\text{s}^{-1}$)
V_{son}	Speed of sound ($\text{m}\cdot\text{s}^{-1}$)
V_0	Scanning speed ($\text{m}\cdot\text{s}^{-1}$)
v_L	Isotherm velocity ($\text{m}\cdot\text{s}^{-1}$)
v_p, m_p, d_p	Velocity ($\text{m}\cdot\text{s}^{-1}$), mass (kg) and diameter (m) of a particle projected in DED
W_D	Work distance in DED (m)
$w^{cd}/w^{l\alpha}$	Solid and liquid phase solute concentrations
w_0	Chemical composition of the initial alloy
$w_E^{\alpha\beta}$	Eutectic composition
X^0, X^{eq}	Initial fraction and phase fraction

Greek abbreviations

α	Extinction coefficient (m^{-1})
α, β	Names of the present phases

β_v	Volume expansion coefficient (K^{-1})
γ	Surface tension (N/m)
$\gamma_{l/g}$	Liquid/gas interface voltage (N/m)
ε	Emissivity
ε_0	Permittivity of vacuum ($8.854 \times 10^{-12} \text{ A}^2 \cdot \text{s}^4 \cdot \text{kg}^{-1} \cdot \text{m}^{-3}$)
$\dot{\varepsilon}$	Strain rate tensor
$\phi =$	Level set function
φ_{source}	Surface heat flux ($\text{W} \cdot \text{m}^{-2}$)
$\gamma_{s/l}$	Solid/liquid surface tension ($\text{N} \cdot \text{m}^{-1}$)
$\Gamma^{\alpha l}$	Gibbs–Thomson coefficient ($\text{K} \cdot \text{m}^{-1}$)
κ	Curvature of the liquid/gas interface (m^{-1})
μ	Dynamic viscosity ($\text{Pa} \cdot \text{s}$)
μ_k	Attachment coefficient ($\text{m} \cdot \text{s}^{-1} \cdot \text{K}^{-1}$)
μ_m	Magnetic permeability of the medium ($\text{H} \cdot \text{m}^{-1}$)
η_p	Powder/melting area mass efficiency (DED)
Δh	Manufactured layer height (DED and E-PBF) (m)
ΔS_f	Molar enthalpy of melting ($\text{J} \cdot \text{mol}^{-1}$)
$\Delta T_d^{\alpha l}$	Growth supercooling of α phase dendritic structure (K)
$\Delta T_e^{\alpha \beta l}$	Growth supercooling of the eutectic structure of phases ($\alpha + \beta$) (K)
ΔT_0	Solidification interval (K)
ΔZ_{plate}	Height of manufactured layer in L-PBF (m)

ΔZ_{powder}	Powder height in L-PBF $\sim 2 \Delta Z_{\text{plate}}$ (m)
δ	Thermal diffusion depth or chemical diffusion thickness (m)
λ	Laser wavelength (m)
$\lambda_1, \lambda_2, \lambda_e$	Primary and secondary interdendritic spacing and eutectic interlamellar spacing (m)
ν	Poisson's ratio (-)
ω	Void ratio = porosity of the powder bed (-)
ρ	Density ($\text{kg} \cdot \text{m}^{-3}$)
ρ_{fil}	Wire density in WAAM ($\text{kg} \cdot \text{m}^{-3}$)
ρ_{gaz}	Gas density ($\text{kg} \cdot \text{m}^{-3}$)
ρ_p	Particle density ($\text{kg} \cdot \text{m}^{-3}$)
σ	Cauchy stress tensor (Pa)
σ_{SB}	Boltzmann constant (radiative losses) $= 5.67 \times 10^{-8} \text{ W} \cdot \text{m}^{-2} \cdot \text{K}^{-4}$
σ_{th}	Thermal expansion coefficient
σ_Y	Yield stress (Pa)
σ_0	Electrical conductivity ($\text{S} \cdot \text{m}^{-1}$)
τ	Laser-matter interaction time (s)
τ_R	Overlap ratio between beads (L-PBF)
τ_s	Solidification time
θ	Anchor angle ($^\circ$)
θ_s	Orientation of the solidification structure

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